TRANSACTIONS

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CONTENTS

The object of the Society shall be to promote the arts and sciences connected with either the manufacture or treatment of metals, or both.—Constitution A. S. S. T., Art. II.

MANUFACTURE, PROPERTIES, AND USES OF 18-8 CHROMIUM-NICKEL STEEL WIRE

By W. H. Wills and J. K. Findley

Abstract

Wire products constitute about 9 per cent of our national output of steel. The tonnage of alloy steel wire is comparatively small due to the excellent physical properties that are obtainable in carbon steel wire. The advent of the high chromium corrosion resisting steels has resulted in a valuable and growing addition to the alloy steel wire division. Of these the 18 per cent chromium-8 per cent nickel (18-8 type) shows the best physical properties and general resistance to corrosion.

After a brief outline of the essential points included in wire drawing, a description of the process for "18-8" wire is given. Following this is a discussion of the physical properties of the wire and of various corrosion tests. Mention is made of the principal applications.

While there have been cases of failure of the wire to come up to expectations, these have been for the most part confined to the hard drawn material and due to lack of knowledge of the properties. With its many good properties, the prospects are for an increasing number of applications and volume of output of this alloy steel wire.

AMONG the various forms of cold working perhaps none are as important and from a metallurgical standpoint as interesting as wire drawing. In ancient times wire was made by hammering the metal out into a flat sheet and then cutting this into narrow and comparatively short lengths. It is not definitely known who first drew wire through a die instead of hammering it, but it is believed to date back about 1000 years. Wire drawing was an established industry in Paris in the 13th century and in England it started about 100 years later. The substitution of rolled for hammered rods greatly facilitated the process of wire drawing and it was only a little over 50 years ago that continuous wire drawing machines were evolved.

At the present time the tonnage represented by wire and wire

products includes about 9 per cent of the national output of steel products. In some forms wire products represent the finest finished product of the steel makers' art.

The process of wire drawing depends on the ductility of a metal or the capacity for plastic flow resulting not so much from tension as from pressure of the die. This cannot be carried on indefinitely, because with every pass through the die the ductility decreases and the hardness increases and before the breakdown is reached the metal must be annealed to bring back its ductility after which the process may be continued. The drawing of steel into wire is often regarded merely as a process of shaping the material accurately to size, while its effect of greatly enhancing the physical properties is overlooked. The former conception applies to many forms of mild steel wire in particular, but the latter is of utmost importance with such products as rope, spring, music, and aircraft wire. Music wire is an example, it being produced with a tensile strength of 350,000 pounds per square inch or more, or six times that of ordinary structural steel and with a high degree of toughness.

In the field of tool steel wire which includes carbon, high speed and other analyses, the problems involved are not so much concerned with high physical properties as holding to a close size tolerance and developing a degree of hardness or structure that will machine readily. Thus where common soft steel wire is drawn to plus or minus 0.002 inch the usual limit on tool steel is only one-quarter as large and on smaller sizes closer than this. Annealing cycles are longer, necessitating more care in avoiding decarburization.

The amount of alloy steel wire production is rather small mainly because of the excellent physical properties that can be developed by the intelligent cold working of carbon steel wire. With the advent of the high chromium corrosion resisting steels a valuable addition was made to the alloy steel wire division and the prospects are that it will grow materially as time goes on. These steels show good physical properties and at the same time they are suitable for many applications where corrosion has to be considered. Formerly this meant the use of plated steel or one of the nonferrous metals or alloys.

Developments thus far show that among the high chromium stainless steels the 18 per cent chromium-8 per cent nickel (18-8) type shows the best general resistance to corrosion as well as good physical properties. Much information has appeared in the technical press re-

garding the nature and properties of this alloy steel, but the essential points will be recounted briefly.   The general analysis is

	Per Cent
Carbon	0.16 maximum
Manganese	0.60 maximum
Phosphorus and Sulphur	0.030 maximum
Silicon	0.75 maximum
Chromium	16.50 to 20.00
Nickel	7.00 to 10.50

Material worked into wire is held somewhat closer, namely

	Per Cent
Carbon	0.07 maximum
Manganese	0.30 to 0.50
Phosphorus and Sulphur	0.025
Silicon	0.20 to 0.40
Chromium	17.00 to 18.00
Nickel	9.00 to 10.00

Carbon and silicon have some influence on the physical properties of the wire, more so in proportion than manganese, chromium or nickel.   Higher carbon and silicon wire will show a little higher tensile strength.   Silicon much over 0.50 per cent is apt to cause embrittlement or fatigue failure when the material is hard drawn.

In the soft state the steel is fully austenitic, nonmagnetic and cannot be hardened by any heat treatment.   The austenite is unstable, but persists as such due to the sluggishness toward any change involving recrystallization or diffusion.   In a great many applications, this instability is not apparent and does not need to be considered.   However, when subjected to elevated temperatures such as from 800 to 1500 degrees Fahr. or when deformed by cold working, some decomposition of the austenite takes place and resistance to corrosion is somewhat impaired.

When subject to cold working it rapidly develops increased hardness making frequent heat treating necessary.   It is possible to soften the alloy by rapid cooling from temperatures ranging from 1650 up to as high as 2200 degrees Fahr.   However, below 1725 degrees Fahr. the austenite grains are not well developed and more time is required. Temperatures above 1900 degrees Fahr. are unnecessary and are apt to cause deep pitting of the surface unless an atmosphere completely free from oxygen is maintained.   Photomicrographs of 18-8 under various conditions of treatment are shown in Figs. 1 to 6.
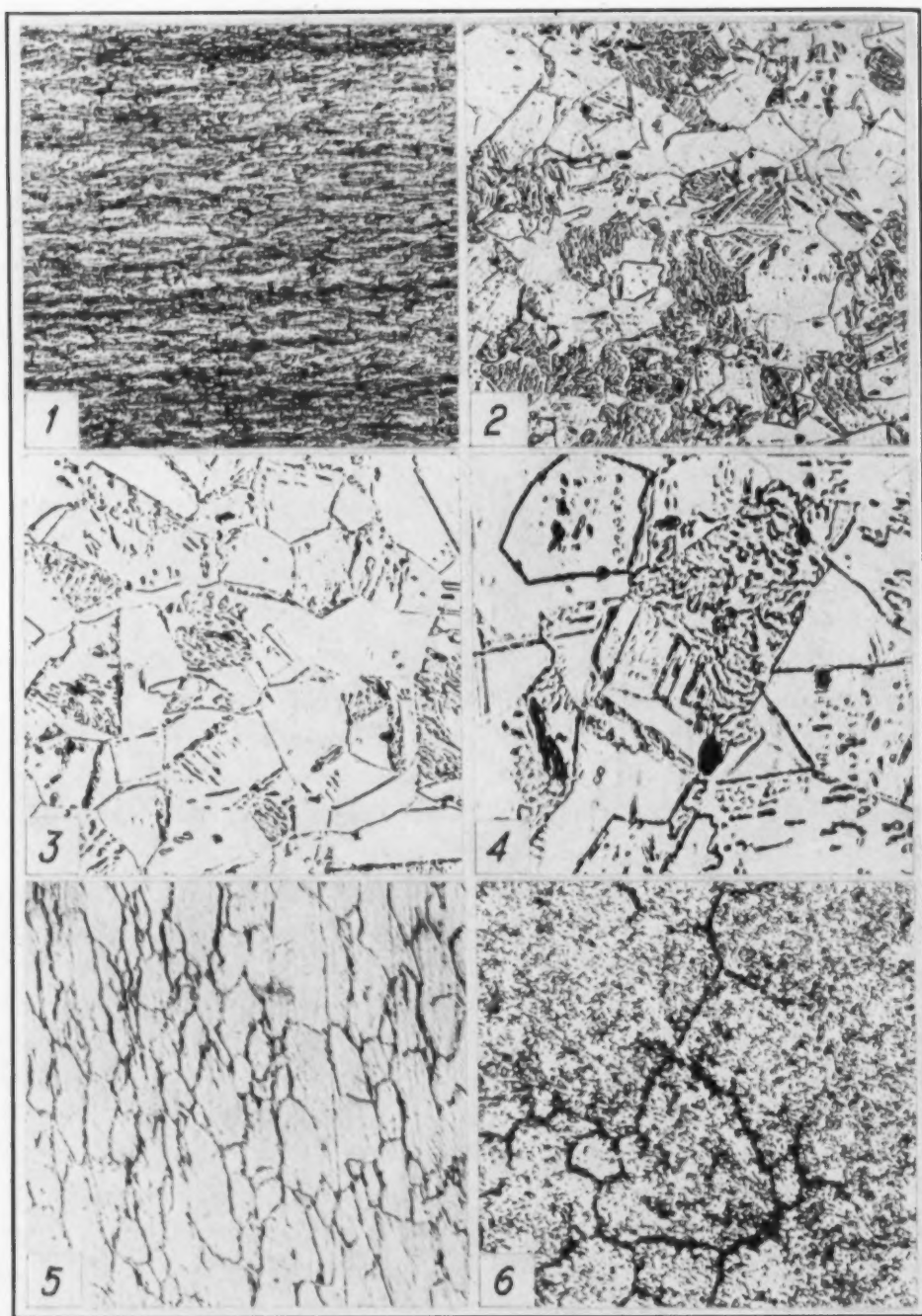
Fig. 1—Structure of ¼-Inch Round Wire Rod (18:8). As rolled condition. × 200.
Fig. 2—Structure of 0.250-Inch Round (18:8) Annealed 1900 Degrees Fahr. Drawn to 0.228-Inch. Then Annealed 1700 Degrees Fahr. Holding 30 Minutes at Temperature, Water Quenched. × 200.
Fig. 3—Same as (2) but Second Anneal at 1900 Degrees Fahr. × 200.
Fig. 4—Same as (2) but Second Anneal at 2100 Degrees Fahr. Note Large Grain Size. × 200.
Fig. 5—Structure of 18:8 Wire After 38 Per Cent Draft from 0.250-Inch. × 200.
Fig. 6—Structure of Hard Drawn 0.028-Inch (18:8) Wire That Had Been Subjected to Severe Corrosive Action. Intercrystalline Corrosion Has Set In. × 1200.

Having such characteristics it is not surprising that some unusual problems are presented in drawing this material into wire. This discussion which is of a practical nature will describe the process of manufacture of 18-8 wire and later touch on the properties and uses. For the benefit of those not familiar with wire drawing it might be well to start with a brief outline of the procedure followed in handling alloy and carbon steel wire.

The raw material of the wire mill consists of hot-rolled wire rods in coils usually of sizes $\frac{3}{16}$ up to $\frac{1}{2}$-inch diameter, although a large tonnage is rolled approximately 0.235-inch round. The weight of the coils varies from as low as 15 or 20 pounds and a usual range of 50 to 75 pounds, in the case of highly alloyed steels, up to 150 to 300 pounds with carbon steels. Owing to the denser nature of most alloy steels lighter drafts are necessary and higher finish rolling temperatures. This of course limits the weight of coil that can be handled and a $\frac{1}{4}$ inch round section is about as small as is practical. Wire rods are rolled on three types of mills. The first of these is the ordinary bar mill type in which the stands of rolls are in one line all coupled together and the rather short billets go back and forth through the various passes by manual handling. This type is now obsolete so far as soft steel is concerned, but is common to the so-called tool steel mills whose products are largely highly alloyed steels with short runs on a size and frequent roll changes. The other two later types are the Belgian or looping mill and the Morgan continuous mill which handle the greater portion of soft steel wire rods. With these mills much larger production is possible with reduced labor costs.

After the coils come to the wire mill, high carbon or alloy material goes through an annealing operation. This is not required with ordinary low-carbon steel wire which in the as-rolled condition is comparatively soft. Annealing of coils before drawing is usually carried out in an open fire. After the coils are drawn the hardness is increased to a greater or less extent depending on the composition. Highly alloyed steels are subject to this work hardening to a considerable extent. When the coils are annealed between passes or after drawing the surface must be protected from scaling. This is accomplished by box or molten bath annealing, or by the use of furnaces with controlled atmosphere.

The hot-rolled rod is completely covered with a layer of scale or iron oxide which after annealing adheres more or less loosely. Before

drawing this must all be removed and this is done chemically by pickling in an acid solution that is heated by steam. Sulphuric acid and muriatic acids are commonly used, the concentration depending on the material going through. For straight carbon wire a comparatively dilute solution of sulphuric acid (4 or 5 per cent) is used while with the high chromium stainless steels greater strength is required. The operation of pickling is very important and much care is necessary in watching the concentration and time of the coils in the bath to avoid the attack of the acid on the metal which may result in pitting of the surface. The addition of inhibitors to the pickling bath has been of great help in avoiding this condition. These are compounds of an organic nature that protect the metal from attack by forming a film over it after the acid has removed the scale. After pickling the coils are thoroughly washed with water and afterwards dipped in a lime water solution which neutralizes the action of the acid and also prevents rusting.

The pickling operation results in the occlusion of hydrogen by the steel, the amount depending on the solubility of steel in the acid, also the temperature of the bath. This renders the wire more or less brittle. However, this is not permanent and the ductility is restored by baking the coils at a temperature of about 400 degrees Fahr.

Before wire can be drawn one end of each coil must be pointed so that it can be started through the die. This is generally done by cold swaging or rolling.

Wire is drawn by pulling the rod through a tapered hole in the die. The least diameter of the hole is smaller than the rod by the amount the stock is to be reduced in a single pass. This of course varies with the kind of steel and its ability to work harden. The simplest form of wire drawing machine consists of a bench on which is mounted a die holder and drum or block which for the larger gages of wire is about 28 inches in diameter. The block which tapers upward rotates on a vertical spindle connected through the bench to a line shaft by bevel gearing. The pointed end of the coil is put through the die and is gripped by a pair of jaws which slowly pulls a sufficient length through the die to go around the block. This is checked for gage and if satisfactory the end is released from the jaws and clamped to the upper edge of the block, which is thrown in gear and draws the remainder of the coil through the die. When the wire passes through one hole only going onto the block the operation is

known as single holing and is the rule with the heavier gages of wire or when handling short coils. The so-called continuous process is carried out with coils of great length, the wire passing through several dies before winding on the block. For drawing the finer sizes of wire, machines of different design are used and as many as 22 dies may be used in series.

The drawing of wire requires considerable power running as high as 50 horsepower for a single die on large units. Lubrication of the wire as it enters the die is consequently an important factor. Whether the wire is coated or not, friction between wire and die is so great that a solid lubricant is in general required. In dry wire drawing tallow or soap in some form is used, while with wet drawing a soluble oil is used. By far the greater part of our output of wire is dry drawn. The higher grades of wire in finer sizes are wet drawn, that is, the dies are submerged under the liquid lubricant. Wire drawn by this process has a finer surface finish and is tougher. Since lighter drafts are taken it is of course more expensive than dry drawing.

Hard wire which is drawn through the dies several times must have a coat on the surface which acts as a lubricant and prevents direct contact between the die surface and the wire. On ordinary material this is simply hydrated oxide of iron and is formed by allowing the wire to wet rust after pickling and before dipping in the lime water bath. Wire of some of the harder steels such as high speed steel is copper coated by dipping the coils in a dilute solution of copper sulphate.

The question of the reduction per pass or system of drafting is an interesting and important one. No hard and fast rules can be given and much depends on the kind of material and how it hardens with the drawing operation. With mild steel typical reduction of area per pass runs from 25 to 45 per cent. On patented high tensile wire this will run 20 to 40 per cent. The average is around 30 per cent on carbon tool steel wire and 15 per cent on high speed steel, a difficult material to draw.

The speed at which wire is drawn depends on the nature of the material. For ordinary mild steel it ranges from 50 to 250 feet per minute for rods $\frac{1}{2}$ to $\frac{1}{4}$-inch diameter and from 250 to 500 feet for wire of size 0.212 down to 0.036 inch. Patented steel is drawn somewhat slower averaging about 30 per cent less.

Drawing dies and their maintenance constitute a vital considera-

tion in every wire mill affecting as they do the cost as well as the quality of the finished product. Chilled cast iron dies are still used in drawing soft steel wire. As they are worn they are reamed to larger sizes. Some mills have replaced chilled iron with high carbon high chromium steel dies. The application of tungsten carbide to wire drawing dies is now developing rapidly and it is only a question of time until they replace other material largely on standard sizes.*

The Process

Having mentioned the essential points of the wire drawing process, its application to the 18-8 chromium-nickel steel will follow. The coils are rolled on the ordinary type bar mill and on the Morgan continuous mill. The use of the latter on this material has not been altogether successful with some producers, due to the low yield of good product. Coils weighing 170 pounds carrying wire about 0.210-inch in size have been turned out on a Morgan mill.

With a 7-stand 10-inch bar mill operating at 240 revolutions per minute it is possible to roll 50-pound coils of ¼-inch material and hold to a tolerance of plus or minus 0.006-inch. By the use of a separate breakdown stand reducing the billets from 2¼ to about ⅞-inch square it would be possible to speed up the mill considerably and roll coils at least 50 per cent heavier.

These 18-8 coils in the as-rolled condition show a hardness of Rockwell C 28 to 30 or approximately 255 Brinell. The first step is an annealing operation carried out in an electric furnace in which a nonscaling atmosphere can be maintained by the introduction of hydrogen or nitrogen. A charge consists of about 800 pounds of ¼-inch wire coils and the total time of the annealing cycle runs around 3½ hours. After heating through at 1900 to 1950 degrees Fahr. (½ hour at temperature) the charge is withdrawn and coils quickly quenched in water. This produces a completely austenitic structure, and the hardness is reduced to Rockwell B 66 to 70 (120 Brinell).

The coils are now ready for pickling and cleaning and about a dozen at a time are immersed in a hot bath of sulphuric acid showing around 10 per cent acid strength. The time in this bath is usually about an hour after it starts to boil. On removing they are thor-

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*Diamond dies are commonly used in drawing very small sizes of wire such as under 0.025 inch.

oughly washed with water and dipped in lime water, after which they are baked for a couple of hours at a low temperature.

This prepares the stock for the first cold drawing operation, the first three passes being known as the roughing passes. This may be carried out single holing, or with continuous drawing, but the economy of the latter cannot very well be realized unless fairly heavy coils are handled. On these passes the draft averages about 0.030-inch or close to 30 per cent reduction per pass, and the stock is drawn through high carbon high chromium steel dies. These steel dies are used down as small as 0.060-inch. Tungsten carbide dies are used from this size down to the very fine sizes which include under 0.025-inch. These dies are showing very good results and it is likely only a question of time until they will be used for larger sizes as well. In fact some producers now use them for the roughing passes as well as the smaller sizes.

After three passes another anneal is required which is carried out similar to the first but at about 1700 to 1750 degrees Fahr. The reason for the lower temperature is that it is sufficient to develop an austenitic structure and there is considerably less scaling which facilitates pickling. Pickling, lime coating and baking are then carried out similar to the corresponding initial treatment.

Two cold passes of about 35 per cent reduction then follow, bringing the size from 0.150 down to 0.100-inch after which there is another anneal at 1700 to 1750 degrees Fahr. Another similar cycle brings the size down to 0.060-inch.

The 0.060-inch stock after getting a 1700-degree Fahr. anneal is given a pickle in a strong muriatic acid solution at 160 to 180 degrees Fahr. On these smaller sizes muriatic acid has been found more efficient than sulphuric. This is followed by baking and lime coating and two passes through tungsten carbide dies brings the size to 0.038-inch. Another cycle like this reduces the size to 0.031-inch.

The coils are now given several passes through tungsten carbide dies in series on a Darmstadt type multiple drawing machine and the size becomes approximately 0.020-inch.

After an anneal at 1700 to 1750 degrees Fahr. followed by a muriatic acid pickle and dip in a nitric acid solution coils are washed and dried and now have what is called a bright pickled finish. This final pickling is followed whenever orders specify this kind of a finish.

The stock at this size is now ready for drawing into the so-called

fine wire sizes. All of this material is wet drawn on multiple die machines using diamond dies. The sizes commonly drawn range from 0.018 down to 0.004-inch though smaller than this can be drawn. The number of dies in series on a machine is usually about 12.

If soft bright wire is ordered it is annealed by passing the wire single strand through a tubular bright annealing furnace held at a temperature of 2000 to 2050 degrees Fahr. and in which a hydrogen atmosphere is maintained. The higher annealing temperature is required in this continuous annealing as the wire travels rather rapidly through the furnace and the time at temperature is very short. Best corrosion resisting properties are brought out with an annealing range of 1950 to 2000 degrees Fahr. This method of annealing is used when necessary to anneal between passes in the fine sizes.

PROPERTIES AND USES

With this picture of the process in mind, let us consider the physical properties shown by this alloy steel wire in its various forms and further on call attention to some of the more important applications.

In the soft condition 18-8 wire shows practically the same physical properties as those of annealed sheets, strips, tubes, and bars and typical figures for sizes $\frac{1}{4}$ down to $\frac{1}{16}$ inch will show:

Yield Point	40,000 pounds per square inch
Ultimate Tensile Strength	90,000 pounds per square inch
Elongation in 8 Inches	55 per cent
Reduction of Area	75 per cent
Rockwell B Hardness	66 to 70

Smaller sizes under $\frac{1}{16}$ inch will show:

Yield Point	40,000 to 55,000 pounds per square inch
Ultimate Tensile	90,000 to 115,000 pounds per square inch
Elongation in 8 Inches	30 to 50 per cent
Reduction of Area	70 to 75 per cent

As the annealed wire is drawn, considerable change is brought about in these physical properties. The austenitic grains are elongated accompanied by precipitation of carbides and formation of ferrite. The yield point rises rapidly with the reduction or per cent of draft, and with as low as 30 per cent approaches within about 10,000 pounds of the ultimate tensile strength. The latter likewise goes up rapidly with increasing per cent draft. Fig. 7. Starting at about

90,000 pounds per square inch it goes up to 150,000 with 30 per cent and to 250,000 with about 75 per cent draft. The ultimate tensile–per cent reduction curve is practically a straight line up to this value. From 75 to 85 per cent draft however, there is pronounced ir-regularity. Drafts above 85 per cent have been made with sizes
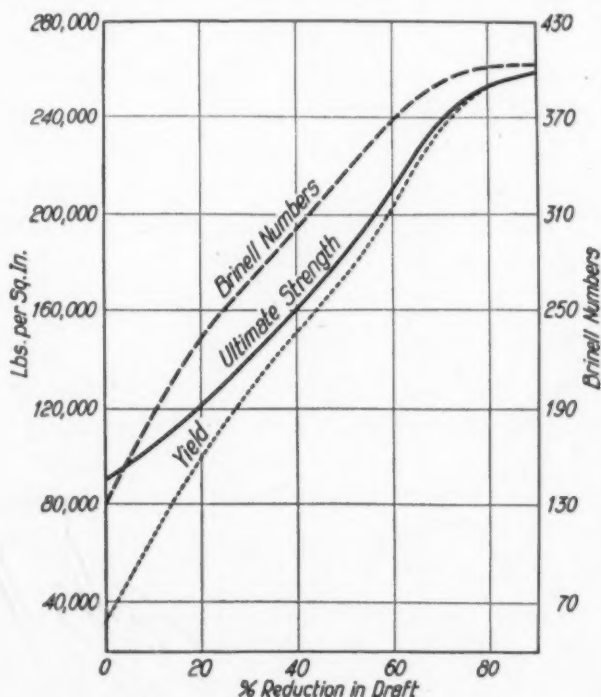
Fig. 7—Curves Showing Relations Between Ulti-mate Strength, Yield Point and Brinell Hardness Numbers of 18:8 and Percentage of Draft from the Annealed Condition. Determinations Made on An-nealed 0.250-Inch Round Stock Drawn to 0.105-Inch.

under 0.030-inch and a tensile strength of 408,000 pounds per square inch was obtained on 0.029-inch wire with a 97 per cent draft. How-ever, 75 per cent is about the maximum practical draft.

Elongation decreases with increasing draft, dropping to 30 per cent in 8 inches at about 25 per cent draft and down to 1 per cent after 60 per cent reduction. Reduction of area, on the other hand, drops more slowly, namely to about 55 per cent after 60 per cent draft. Hardness shows rapid increase. For example Rockwell B 66 to 70 (120 Brinell) annealed to Rockwell C 20 to 23 (235 Brinell) with 30 per cent draft increases to Rockwell 46 to 47 (460 Brinell) after 70 per cent reduction. Above this amount the hardness shows little change.

The ratio of the Brinell hardness numeral divided by the tons per square inch ultimate strength for annealed 18-8 wire is about 3.2 or somewhat lower than for ordinary mild steel. However, after 15 to 20 per cent draft this rises to about 4.2 which is close to the corresponding ratio for medium carbon steel. The value drops to 3.5 at 250,000 pounds per square inch strength following very much the figure for mild steel.

The torsional strength of 18-8 wire is approximately 75 per cent of the ultimate tensile strength. Good high tensile carbon wire will show superior torsion values as compared with hard drawn 18-8 wire. The former will show a more consistent number of turns before rupture on this test whereas the latter tends to be rather erratic in this respect.

The electrical resistance of the annealed 18-8 wire is somewhat lower than the hard drawn. This constant is 73 microhms per cubic centimeter, while 250,000-pound tensile hard-drawn will show 96 microhms per cubic centimeter. The wire is entirely nonmagnetic when softer but becomes magnetic upon cold drawing. This is pronounced when the draft has been sufficient to produce a tensile strength of 185,000 to 215,000 pounds per square inch or a reduction of around 50 per cent.

General resistance to corrosion of the annealed wire is similar in all respects to that shown by other forms of this material in the fully soft condition. The resistance to such agents as salt water, fruit juices, chlorine or sulphur dioxide bleaching solution, and dilute nitric acid is practically unaffected by cold drawing. Hard drawn wire is somewhat more subject to attack by strong mineral acids, strong alkalis, and moist hydrogen sulphide. However, the resistance of the hard drawn wire to these strong corrosive agents is much greater than that of any of the straight chromium alloy types.

The accompanying table gives one an idea of how the annealed or fully soft condition compares with medium or hard drawn wire when exposed to various corrosive agents.

As might be inferred from one of the introductory statements, 18-8 wire is in competition with high grade carbon wire that is tinned or otherwise plated to resist corrosion and wire of some of the nonferrous alloys such as monel metal, brass, or phosphor bronze. As compared with the former the difference in physical properties is not marked, but there is the disadvantage of the plating wearing off, par-

Corrosion Test Results
18-8 Wire—Size 0.020-inch

	Percentage Loss in Weight—24 Hours Exposure		
	Annealed 1950 Degrees Fahr.	Medium Hard Drawn 155,000 Pounds per Square Inch	Hard Drawn 251,000 Pounds per Square Inch
25 per cent HNO_3 Boiling	0.00	0.00	0.05
50 per cent HNO_3 Boiling	0.01	0.05	0.05
Conc. HNO_3 Boiling	0.01	0.05	0.05
50 per cent HCL Cold	40.00	75.00	75.00
50 per cent HCL Boiling	100.00	100.00	100.00
Conc. HCL Cold	50.00	75.00	75.00
Conc. HCL Boiling	100.00	100.00	100.00
20 per cent $H_2 SO_4$ Cold	50.00	60.00	65.00
20 per cent $H_2 SO_4$ Boiling	100.00	100.00	100.00
Conc. $H_2 SO_4$ Boiling	100.00	100.00	100.00
8 per cent $H_2 SO_3$ Cold	2.00	14.00	20.00
8 per cent $H_2 SO_3$ Boiling	8.00	100.00	100.00
3 per cent $H_2 SO_3$ Cold	0.08	4.00	12.00
3 per cent $H_2 SO_3$ Boiling	2.00	20.00	30.00
Conc. Acetic Acid Boiling	0.00	0.00	0.00
50 per cent Tartaric Acid Boiling	0.00	0.00	0.00
Moist Hydrogen Sulphide (36 hours)	0.00	0.02	0.10
Saturated Salt Spray (36 hours)	0.00	0.00	0.00
10 per cent $H_2 SO_4$—10 per cent $CuSO_4$	0.00	0.00	0.00
20 per cent Sodium Hydroxide Cold	0.00	0.00	0.00
20 per cent Sodium Hydroxide (200 degrees Fahr.)	0.00	0.01	0.04
Ammonia Water (800 degrees Fahr.)	0.00	0.00	0.00
Live Steam (36 hours)	0.00	0.00	0.00
Air, heavy with smoke containing sulphur (36 hours)	0.00	0.00	0.01

ticularly if subject to abrasion. Making comparison with the latter or nonferrous metal wires there is a decided advantage in physical properties and resistance to abrasion. Fig. 8. However, there is one disadvantage that must be taken into account. That is, the tendency to work harden on applications where the wire is subject to continual bending back and forth.

18-8 wire is being fabricated into wire rope. In this form the cost is roughly five times that of the best carbon steel wire rope. For this purpose such sizes as 0.060 down to 0.015-inch are commonly used and ordered to a tensile strength of about 250,000 pounds per square inch. This product stands salt water corrosion and ordinary weathering and works out well on marine applications.

Perhaps the largest outlet for the wire thus far is for screens and wire cloth. One concern has woven as fine as 0.0045-inch wire into 100-mesh cloth and as coarse as 0.020-inch into 14 mesh. Wire for weaving is usually furnished bright annealed. The bright annealing must be watched carefully for the slightest discolored condition will be readily apparent in the cloth. Also the wire must be free from kinks which will produce roughness of the cloth.

The looms employed in the production of wire cloth resemble in general principles those used in the textile industries, but there are points of difference made necessary by the untractableness of wire as compared with cotton or wool yarns and the speed of operation is less.
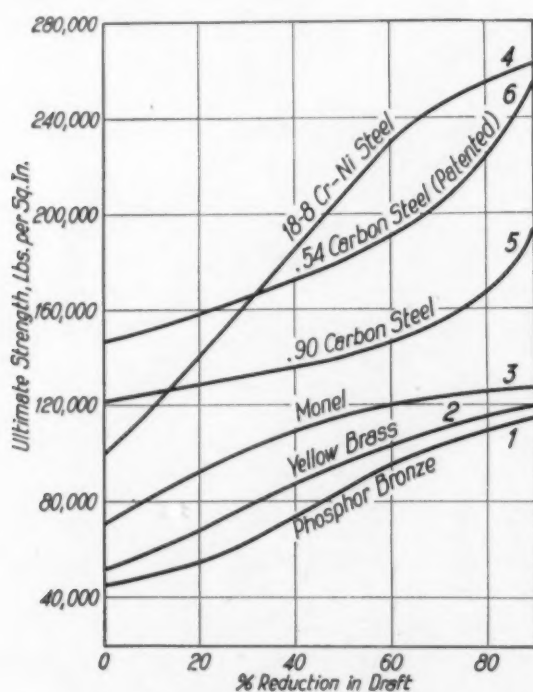
Fig. 8—Comparative Curves Showing Relation Between Ultimate Tensile Strength and Percentage Reduction in Draft After Annealing. Curves 1, 2, 3, 4 and 5 Are for Fully Annealed Material Size 0.020-Inch Cold Drawn to 0.007-Inch. Curve 6 for Patented 0.54 Per Cent Carbon Steel Wire Drawn from 0.192 to 0.051-Inch.

On fine 200-mesh cloth with a 6-foot loom only 4 or 5 feet will be turned out in a working day, while on 40-mesh the production will be 20 to 25 feet.

The greater part of this wire cloth is for stationary or centrifugal filters in the chemical industry and they also find use in paper mills.

Another important application is for welding rod. Welding by gas and particularly by the electric arc has developed to such an extent that it is now common in connection with the fabrication of tubes, plates, and bars, also in salvaging forgings or castings. For best results the welding rod should be of essentially the same composition as the parts to be welded. Thus the output is roughly propor-

tional to that of the other forms of 18-8 produced. Rod of this composition is used with either gas or electric welding. Having no capacity for hardening by the effect of temperature the welds remain tough and ductile and do not develop the coarse crystalline structure that is apt to be encountered with the straight iron-chromium alloys.

In acetylene welding the flame should be held as near neutral as possible with the gases admitted at low pressure. If a reducing flame is used there is danger of carbon pickup. On account of the lower heat conductivity as compared with ordinary steel the torch should be kept pointed in the direction of the work so as to preheat the parts to be welded.

In welding sufficient heat is taken up at or near the welded area to cause carbide precipitation with resultant decrease in resistance to corrosion. Consequently for maximum corrosion resistance all welded parts should be reheated to 1900 to 2000 degrees Fahr. followed by rapid cooling to produce a fully austenitic structure.

This material is somewhat difficult to cold head which is to be expected considering its work hardening property. It is cold-headed successfully where the displacement of metal is moderate and is being worked into rivets and bolts.

There have been quite a number of applications for machine screws for screen plate work in paper mills. These hold the plates in place in the frames and are replacing bronze screws which are attacked by the alum in the water. The 18-8 screws are not attacked and once installed can be used repeatedly for the same size.

18-8 wire has been worked up into nails, but is rather hard on the dies. There is an application for such nails holding drainer box plates in sulphite mills. Also in powder plants, screws and nails are used in the wash tanks where the acid is separated from the nitrated cotton.

In the form of wood screws there are numerous uses for trim work on furniture, dairy equipment, cold storage doors, and building decorations.

It has proved a satisfactory material for automotive wire wheel spokes. For this purpose it is furnished with a yield of about 100,000 pounds per square inch, 30 per cent elongation in 2 inches, 65 per cent reduction of area and the hardness runs Rockwell C 21 to 24. A minimum of six 90-degree bends are required. The wire is worked into nipple type spokes and one end is subject to a light cold-heading operation. Results on the upset type of spoke in which both ends are

cold-worked have not been very successful. Here a steel that will not work harden so much is necessary.

A considerable quantity of the wire finds use for dental work, such as in connection with the straightening of teeth and for certain dental tools.

CONCLUSION

The foregoing uses give one some idea of the progress that is being made by this comparatively new alloy steel wire. Besides these there are many minor applications and new ones that are constantly developing. There have of course been cases of failure of the wire to come up to expectations, this applying mostly to hard drawn material, but this has been due in general to an improper conception of the properties. With its many good qualities, and judging from results obtained in service, the prospects are very good for continued growth in the production of this wire.

In conclusion the writers wish to express their thanks to F. B. Lounsberry, vice-president of Ludlum Steel Co., for permission to present the information and data given in this paper, and for helpful suggestions regarding its preparation.

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DISCUSSION

Written Discussion: By R. C. Jordan—Wickwire Spencer Steel Co., Worcester Mass.

Mr. Wills and Mr. Findley have collected some very useful data and have presented it in an interesting and instructive form.

Considerable data on tensile strength increase with drafting confirms in general the drafting curves No. 4, No. 5 and No. 6 in Fig. 8. If data for a low carbon steel, such as 0.07 per cent to compare directly with the analysis of 18-8, had been shown, it would have indicated the same general curvature downward on the chart as does the 18-8, for drafting beyond 65 per cent reduction of area. This indicates the very marked influence of carbon content on hardening due to extreme cold work.

Accompanying curves show data on the effects of varying annealing temperatures on the tensile strength and elongation. The test was made by drawing

a strand of 18-8, size 0.035 inch, through a furnace 5 feet long at a speed of 35 feet per minute, quenching in air. Temperatures by 50 degrees Fahr. increments are shown on the curve. The wire was protected by nickel tubes, temperature readings being made outside the tubes, hence the actual heat absorbed
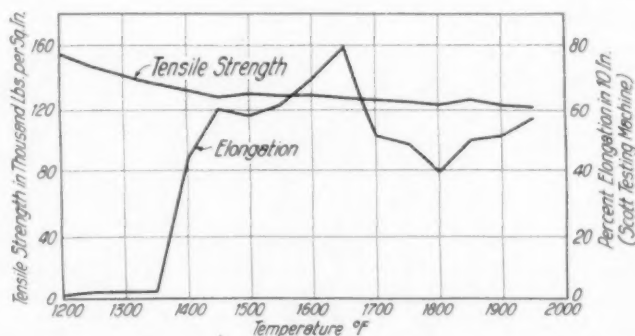
by the wire was probably less than that shown, due to the continuous operation. In any continuous operation of a furnace the temperatures must be considered as relative.

In the plotted data it will be noted that the tensile strength decreases gradually and uniformly, though at no time does it approach the softness of a full anneal. The elongation, however, shows a marked increase at as low as 1400 degrees Fahr. and another very marked maximum at 1650 degrees Fahr., after which a further decrease and increase is noted. Elongation tests were made on a Scott testing machine with a 10-inch gage length, which will give a higher value for elongation than an Olsen or Riehle type tester with recording apparatus actuated by the stretch in the wire instead of by the jaws of the machine holding the test specimen.

A similar curve on another austenitic steel of different analysis showed the same peculiar characteristics of the elongation change.

Any temperature apparently which gives a uniformly good elongation may be used for a process annealing point for this alloy. It is, however, reasonable to conclude that the temperature giving the highest elongation should prove most satisfactory for subsequent drafting. It would be interesting to know whether the authors of this paper have any theory to explain the apparent eccentricities of the elongation with various annealing temperatures.

DR. JOHN A. MATHEWS:[1] I think this is one of the most interesting papers on wire drawing that I have ever seen presented, in fact, papers on that subject have been very scarce.

The point that I wish to mention, however, is in reference to the character of the austenitic steels in general. From some experiments we have made, it would seem to be a characteristic of the customary 18-8 composition to develop more work hardening than any of the other modifications. Raising of nickel in that steel would cause lower tensile strength and work hardening. Chromium irons of course do not work harden to anywhere near the same extent. Possibly a broader generalization is that you might say that the border

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[1]Vice-president, Crucible Steel Company of America, New York City.

line alloys have more capacity for work hardening than the fully austenitic or any of the chromium ferritic steels.

Authors' Closure

The authors wish to thank Mr. Jordan and Dr. Mathews for their discussion. The curves presented by Mr. Jordan do not agree with our curves for tensile strengths and elongations after different annealing temperatures. When the wire is processed as outlined in the article it will be found that the tensile strength drops gradually between the temperatures of 1000 and 1350 degrees Fahr., then drops quite rapidly to about 1700 degrees Fahr., where it starts to
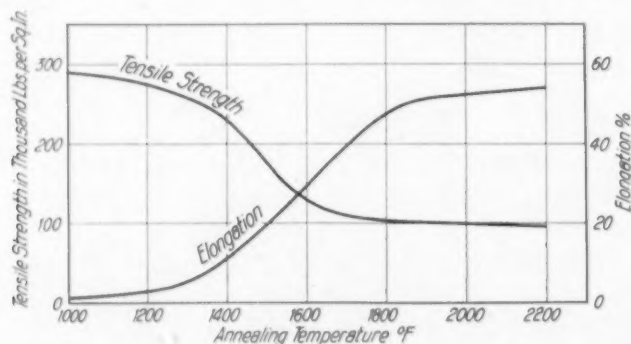
level off and becomes almost constant after 1850 degrees Fahr. The elongations have a rather rapid rise at somewhere around 1300 degrees Fahr. and rise rapidly till the annealing temperature has reached 1850 degrees Fahr., after which there is very little increase in elongation with increasing annealing temperatures.

The accompanying curves are averages for sizes under 0.060 inch. It will be noted that the curves are for (hard drawn) wire showing initial tensile of nearly 300,000 pounds per square inch. Where the initial tensile strengths are lower the curves will be nearly parallel until annealing temperatures between 1300 and 1400 degrees Fahr. are reached and there the curves will converge, meeting somewhere between 1600 and 1725 degrees Fahr. We cannot account for Mr. Jordan's findings unless it is in the treatment of the wire before annealing or that the analysis has varied from that which we use.

We have found that an occasional spool or coil of wire when annealed at a given temperature will not show over, say 45 per cent elongation, whereas the others that have had the same treatment will show elongations of 55 per cent.

When 18-8 is drafted over 92 per cent the tensile strength rises rapidly to over 300,000 pounds per square inch; that is, the tensile strength-percentage draft curve makes a sharp upturn after percentage above 92. As our experience has been with the 18-8 type only we do not know whether the above would apply to other chromium-nickel types.

INFLUENCE OF SPECIAL ELEMENTS ON THE CARBON CONTENT OF THE IRON-CARBON EUTECTOID

By Everett L. Reed

Abstract

The effect of nickel, manganese, chromium, tungsten, molybdenum, vanadium, silicon, copper, aluminum, cobalt, titanium, and uranium on the carbon content of the iron-carbon eutectoid has been studied. The author determined these eutectoid displacements by adding special elements in increments of 1 per cent to pure iron-carbon bases containing respectively 0.16, 0.32, 0.59, and 1.50 per cent carbon and in most cases melted twice in vacuum. These alloys were slowly cooled through the critical range. It was determined whether the alloy was either hypoeutectoid, eutectoid or hypereutectoid by microscopic analysis and in many instances the eutectoid points were confirmed by hardening power determinations.

A study of segregation of carbon and special elements was made in some of the alloy steels as cast.

It was determined that hypereutectoid nickel steels are martensitic and hypereutectoid manganese steels made from medium carbon steel bases are also martensitic. Hypereutectoid chromium, tungsten, molybdenum, and vanadium steels were cementitic. The special elements may be arranged in order according to their effectiveness in displacing the iron-carbon eutectoid as follows: vanadium, manganese, chromium, molybdenum, tungsten and nickel.

Special elements which have a body-centered cubic lattice shift the eutectoid point in the iron-carbon equilibrium diagram to the left while face-centered cubic elements except nickel displace the eutectoid point to the right. The relation of special elements which enlarge and close the gamma field to their influence or displacing the iron-carbon eutectoid is discussed. Brinell hardness of the hypo-

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eutectoid, eutectoid, and hypereutectoid alloys in the cast, annealed and normalized condition has been studied.

A study of the microstructure and hardness of the special eutectoid steels was made after spheroidizing.

The effect of special elements on the critical range was determined.

Physical tests including tensile, Charpy and twisting tests were made on nickel, chromium and molybdenum steels of approximate eutectoid composition. The molybdenum steel possessed the maximum torsional strength at high temperature. The chromium steel is not strengthened in the blue heat range.

Introductory

IT is well known that some of the special elements lower the carbon content of the eutectoid. In addition to the interesting work on this subject previously published by Le Chatelier (1),* Stead (2), Scott (3), Aal (4), Guillet (5), Monypenny (6), Honda (7), and Ishiwara (8), the author presents a systematic study on the influence of twelve special elements on the composition of the iron-carbon eutectoid. The fact that very pure iron-carbon alloys (those containing very low manganese, phosphorus, sulphur and silicon) were used in preparing the alloy eutectoid steels and that these alloys were melted in vacuum should be considered when comparing the results presented in this paper with those of the authors mentioned above.

Scope of the Present Work

The scope of the present paper may be classified as follows:

1. Details of procedure of making the alloys and annealing to promote equilibrium conditions.
2. Determination of the pearlite displacement points by twelve special elements.
3. A study of the microstructure of the annealed hypoeutectoid, eutectoid, and hypereutectoid alloys.
4. Influence of special elements on the hardness of the hypoeutectoid, eutectoid and hypereutectoid alloys.
5. Determination of the hardening power of the alloy eutectoid steels.
6. Thermal analysis of the hypoeutectoid, eutectoid and hypereutectoid steels.

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*Numerals in parentheses pertain to the references appended to this paper.

7. Influence of special elements on spheroidizing of the alloy eutectoid steels.
8. Physical properties of some alloy eutectoid steels.
9. Discussion and conclusions.

THE AUTHOR'S METHOD OF DETERMINING THE EUTECTOID DISPLACEMENT

Charges of pure iron-carbon alloys, containing respectively 0.16, 0.32, 0.59 and 1.50 per cent carbon were prepared to which additions of 1 per cent of pure special element were added and melted in vacuum. In most cases the ingot was inverted in the crucible and melted in vacuum a second time, thereby eliminating the possibility of pronounced segregation of the element.

The melt was heated to 1000 degrees Cent., held at this temperature for one hour and cooled slowly through the critical range, after which the specimen was examined for 100 per cent pearlite. By preparing a series of iron-carbon alloys each containing additions of 1 per cent special element, the author was able to determine whether the alloy was hypoeutectoid, eutectoid or hypereutectoid.

The selection of specimens for microscopic examination was made as follows:

1. One hypoeutectoid alloy (nearest the eutectoid composition) from the open-hearth iron bases containing respectively 0.59, 0.34, and 0.16 per cent carbon.
2. One eutectoid alloy from the open-hearth iron bases containing respectively 0.59, 0.34, and 0.16 per cent carbon.
3. One hypereutectoid alloy (nearest the eutectoid composition) from the open-hearth iron bases containing respectively, 0.59, 0.34, and 0.16 per cent carbon.

The eutectoid displacements were determined by adding special elements to an iron-carbon base, until the alloy was composed of one hundred per cent pearlite. The eutectoid points were confirmed by hardening power determinations. The hardening power was generally found to be maximum at the eutectoid composition.

The difference in hardness between annealed and hardened steel of the same composition is known as the hardening power.

PREPARATION OF THE ALLOYS

1. Preliminary Statements

Oxidation of the melt was prevented by melting in vacuum. Pronounced segregation of the special element was eliminated in most cases by melting the ingot twice. The degree of segregation of carbon and the special element after double melting was studied.

2. Materials Used

(a) *Iron and Iron-Carbon Alloys.* Open-hearth iron (Armco) and pure iron-carbon alloys were used as bases to which the special element was added. Chemical analyses of these alloys are shown in Table I. The manganese content of these alloys is very low, likewise the phosphorus, sulphur and silicon contents are well below the minimum in commercial steels. The silicon content, however, in sample No. 244 compares quite closely with that of commercial crucible steel of tool steel carbon content.

(b) *Special Elements.* A list of elements used in making the alloy eutectoids is given below:

			Per Cent
1.	Nickel (Electrolytic)		99.54
	Impurities—Cobalt		0.40
		Carbon	0.04
		Iron	0.01
		Sulphur	0.001
2.	Manganese Pure, Distilled		
3.	Aluminum		99.6
4.	Cobalt (Electrolytic)		99.8
5.	Copper (Electrolytic)		99.99
6.	Chromium (Electrolytic)		99.41
	Impurities—Hydrogen		0.45
		Iron	0.14
7.	Molybdenum		99.00
8.	Tungsten		99.97
	Impurities—0.03% Non-Metallics		
	Probably Al_2O_3 and other refractory oxide		
9.	Vanadium		74.60
10.	Silicon		97.53
	Impurities—Lead		0.04
		Iron	0.05
		Manganese	0.06
	Balance probably oxygen.		
11.	Titanium		99.9
12.	Uranium		90.30

3. Furnace Used for Melting in Vacuum

An arsem vacuum furnace built by the General Electric Company

and capable of melting 70 to 100 grams of metal was used for making the alloys.

4. Melting Procedure

(a) *Iron-Carbon Alloys Melted in Vacuum.* The iron-carbon alloys (shown in Table I) were cut into suitable small sections and cleaned thoroughly. Charges weighing approximately 75 grams were prepared. At the time of preparing the charge, the proper amount of

Table I

Chemical Composition of Materials Used

No.	Material as received	C	Mn	P	Per Cent S	Si	Cu	Ti
47	Armco Iron (Lot No. 1) Plate	0.02	0.018	0.006	0.035	0.01		
534	Armco Iron (Lot No. 2) Plate	0.02	0.016	0.0025	0.018	Trace		
234	0.16% C, Armco base Cast Ingot	0.16	0.063	0.004	0.027	0.005		
246	0.32% C, Armco base Cast Ingot	0.32	0.063	0.004	0.027	0.01		
247	0.59% C, Armco base Cast Ingot	0.59	0.079	0.005	0.029	0.02		
243	0.83% C, Armco base 1" round rod	0.83	0.07	0.003	0.020	0.04	0.04	0.0025
244	1.13% C, Armco base 1" round rod	1.13	0.08	0.004	0.003	0.17	0.048	
245	1.50% C, Armco base Cast Ingot	1.50	0.007	0.004	0.026	0.05		

the element was weighed on an ashless filter paper and placed in the bottom of the crucible, after which the iron-carbon base was placed in the crucible. The two elements, titanium and manganese, reacted with the walls of the crucible at high temperature. This deleterious effect was greatly eliminated by the following procedure:

A ⅜ inch hole was drilled to a depth of nearly three-quarters of the length of the iron-carbon base called the container (Fig. 1). This hole was threaded in order that a suitably machined cap could be easily screwed into the container resulting in a closed chamber. The entire iron-carbon base, container and screw cap were cleaned, washed in ether to eliminate oil and grease, and weighed preparatory to charging the special element. The element was inserted in the chamber as shown in Fig. 1 and sealed securely by screwing the plug into the container very tightly. The element uranium in addition to the elements manganese and titanium was enclosed in containers previous to melting.

The melt was held at 2200 degrees Fahr. (1205 degrees Cent.)

for ten minutes and cooled rapidly in the vacuum furnace by turning off the power in increments of one kilowatt every 3½ minutes. The ingot was cooled to black heat in thirty minutes.

As previously stated, in most cases the ingots were melted a second time to promote homogeneity. The ingot taken from the
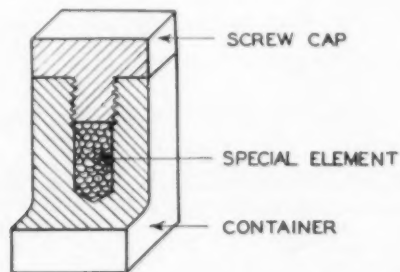
Fig. 1—Cross Section of Metal Container in Which the Special Elements Manganese, Titanium and Uranium were Sealed Previously to Melting.

crucible after the first melting was inverted and placed in a new crucible and remelted according to the practice previously explained.

5. *Annealing the Ingot*

It is obvious that equilibrium conditions are essential in studying the austenite-pearlite transformation. Some special elements make it a sluggish one, therefore in order to approach the ideal case of a complete phase change (gamma to alpha) the author adopted the method of cooling the specimens very slowly in a well heat insulated gas furnace. A brick wall was built around a No. 17 gas furnace (made by the American Gas Furnace Co.) care being taken to provide for two exhausts—one at the top of the furnace and the other at the back of the furnace. The floor of the brick chamber was lined with asbestos transite board. The entire chamber was filled with a mixture of powdered asbestos and powdered sil-o-cel. Under these conditions the furnace was buried in the center of a bed of heat insulating material measuring about four feet square and two and one-half feet deep.

A cooling curve of the furnace was recorded after holding the furnace one hour at 1000 degrees Cent. and cooling from 1000 degrees Cent. to 50 degrees Cent. (a period of sixty hours).

In view of the fact that such slow annealing* would obviously promote severe decarburization and oxidation of the specimens, they were packed in steel tubes containing either cast iron chips or sand. The medium high carbon alloys were embedded in cast iron chips while the alloys of iron and special element and the low carbon alloys were sealed in sand filled containers.

6. *Preparation of the Ingots for Subsequent Examination*

Fig. 2 illustrates the method of cutting the ingots, so that proper
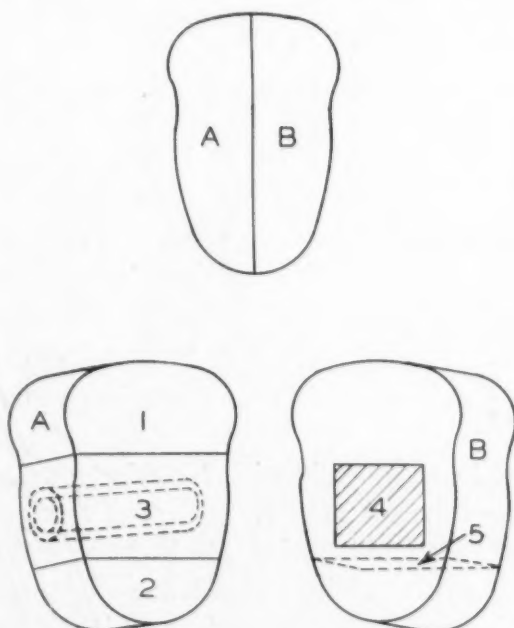
Fig. 2—Preparation of Ingot. Section A, Nos. 1 and 2—Chemical Analysis, No. 3— Thermal Analysis. Section B, No. 4—Micro- scopic Analysis Brinell Hardness Tests. No. 5 —Spheroidizing Tests.

test specimens could be obtained for studying the constitution of the alloys.

7. *Chemical Analysis*

The annealed longitudinal section of the ingot was cleaned thoroughly and clamped in the vise of a milling machine. The milling cutter, revolving at a slow speed, milled off fine chips which were sat-

*Reference to "annealed specimens" refers to samples which have been held one hour at 1000 degrees Cent. and cooled in tubes packed with either cast iron chips or sand.

isfactory for analysis. As shown in Fig. 2, samples for chemical analysis were taken from the upper and lower sections of the ingot.

THE INFLUENCE OF NICKEL ON THE COMPOSITION OF IRON-CARBON EUTECTOID

The eutectoid displacements determined by microscopic examination and hardening power determinations are plotted in Fig. 3. The author's results agree closely with those of Scott (to a maximum of 10 per cent nickel) but are appreciably lower than those of Aal. The results of Scott (10) and Aal (11), are hypothetical in excess of 10 per cent nickel.

The following eutectoid points are calculated from the author's curve shown in Fig. 3:

Per cent carbon	Per cent nickel necessary to produce 100 per cent eutectoid
0.1	13.20
0.2	11.95
0.3	11.10
0.4	10.20
0.5	8.70
0.6	6.50
0.7	4.35
0.8	2.50
0.9	0

The annealed hypoeutectoid nickel steels contained free ferrite and sorbito-pearlite (Figs. 4 and 5).

The annealed eutectoid nickel steels have either pearlitic or sorbito-pearlitic structures (Fig. 6) with the exception of the eutectoid alloy containing 0.07 per cent carbon and 13.48 per cent nickel which contains massive cementite (Fig. 7).

It was found that annealed hypereutectoid nickel alloys were martensitic. No cementite was found in the microstructure. It is evident therefore that as nickel is progressively added to a hypoeutectoid steel, the ferrite disappears, the eutectoid is formed and then the alloy becomes martensitic (hypereutectoid)—(Fig. 8).

Microstructure of Annealed Iron-Nickel Alloys

(a) *Hypoeutectoid—0.03 per cent carbon, 13.56 per cent nickel.* The presence of two phases (12) are evident in Fig. 9. The dark constituent is probably an aggregate of the nickel-rich and the iron-rich phases while the white areas may represent an excess of the alpha

iron-rich phase. The white areas increase in the microstructure with the decrease in nickel.

(b) *Eutectoid—0.03 per cent carbon, 14.76 per cent nickel.* Equal mixtures of the two alpha and gamma phases are present as shown in Fig. 10.

(c) *Hypereutectoid—0.04 per cent carbon, 15.58 per cent*
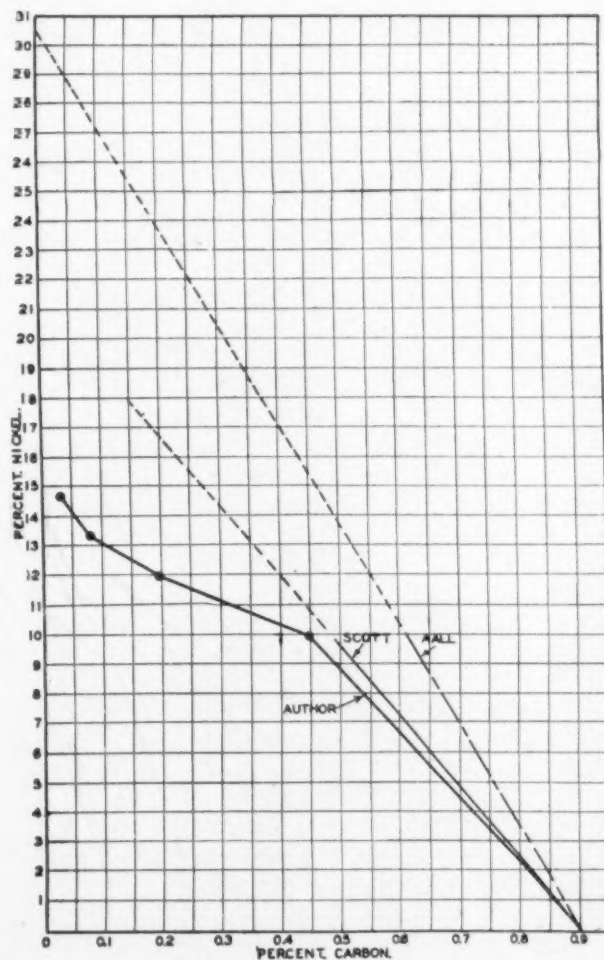
Fig. 3—Influence of Nickel on the Carbon Content of the Iron-Carbon Eutectoid as Determined by Aal, Scott and the Author.

nickel. With further additions of nickel an increase in the proportion of the white constituent was found which is probably the gamma phase rich in nickel. The resulting microstructure consists of the excess gamma phase and an aggregate of the alpha and gamma phases.

The chemical analysis of the series of iron-carbon-nickel alloys

Table II
Composition of Iron-Carbon-Nickel Alloys

| | Calculated Charge | | | Chemical Analysis Melted Twice in Vacuum | |
| | Armco Iron-Carbon Alloys Containing per cent Carbon | Per cent Nickel added | Desired | *Per cent combined carbon | Per cent Nickel |
No.					
135	0.59	8.9	hypoeutectoid	0.41	8.82
134	0.59	8.9	eutectoid	0.45	9.79
136	0.59	10.9	hypereutectoid	0.38	10.78
158	0.32	12.0	hypoeutectoid	0.11	10.36
145	0.32	13.0	eutectoid	0.20	11.90
144	0.32	14.0	hypereutectoid	0.22	13.57
179	0.16	13.0	hypoeutectoid	0.10	13.09
160	0.16	14.0	eutectoid	0.07	13.48
159	0.16	15.0	hypereutectoid	0.05	15.03
167	Armco iron	13.5	hypoeutectoid?	0.03	13.56
181	Armco iron	14.5	eutectoid?	0.03	14.76
180	Armco iron	15.5	hypereutectoid?	0.04	15.58

*No graphitic carbon found in these alloys.

is shown in Table II. In most cases there was no appreciable loss of nickel by double melting in vacuum; the loss of carbon, on the other hand, was quite pronounced.

Vacuum-Melted Versus Air-Melted Iron-Carbon-Nickel Eutectoid

Fig. 11 illustrates the microstructure of an air-melted annealed eutectoid iron-carbon-nickel alloy, containing 0.42 per cent carbon and 9.65 per cent nickel.

Let us return to Fig. 6 which shows the microstructure of a vacuum-melted annealed eutectoid iron-carbon-nickel alloy containing 0.45 per cent carbon and 9.79 per cent nickel. Segregation of nickel and possibly some carbon is present in the dendritic fillings which promote martensitic areas in the matrix as shown in Fig. 11. It appears, however, that the lamellae in the vacuum-melted eutectoid nickel alloy are more pronounced than in the air-melted alloy. It appears from this experiment that oxygen decreases to a slight degree the amount of nickel necessary to displace the eutectoid point. Moreover, we might assume also that oxygen prevents a complete austenito-pearlite transformation.

The Influence of Manganese on the Iron-Carbon-Nickel Eutectoid

An analysis of a commercial steel which was obtained for this study is given below:

Carbon	Manganese	Phosphorus	Sulphur	Silicon
0.66%	0.69%	0.014%	0.031%	0.19%

Fig. 4—Melt No. 135, Hypoeutectoid, 0.41 Per Cent Carbon, 8.82 Per Cent Nickel. As Cast, Annealed. Brinell 212. × 1000.

Fig. 5—Melt No. 179, Hypoeutectoid, 0.10 Per Cent Carbon, 13.09 Per Cent Nickel. As Cast, Annealed. Brinell 262. × 1000.

Fig. 6—Melt No. 134, Eutectoid 0.45 Per Cent Carbon, 9.79 Per Cent Nickel. As Cast, Annealed. Brinell 255. × 1000.

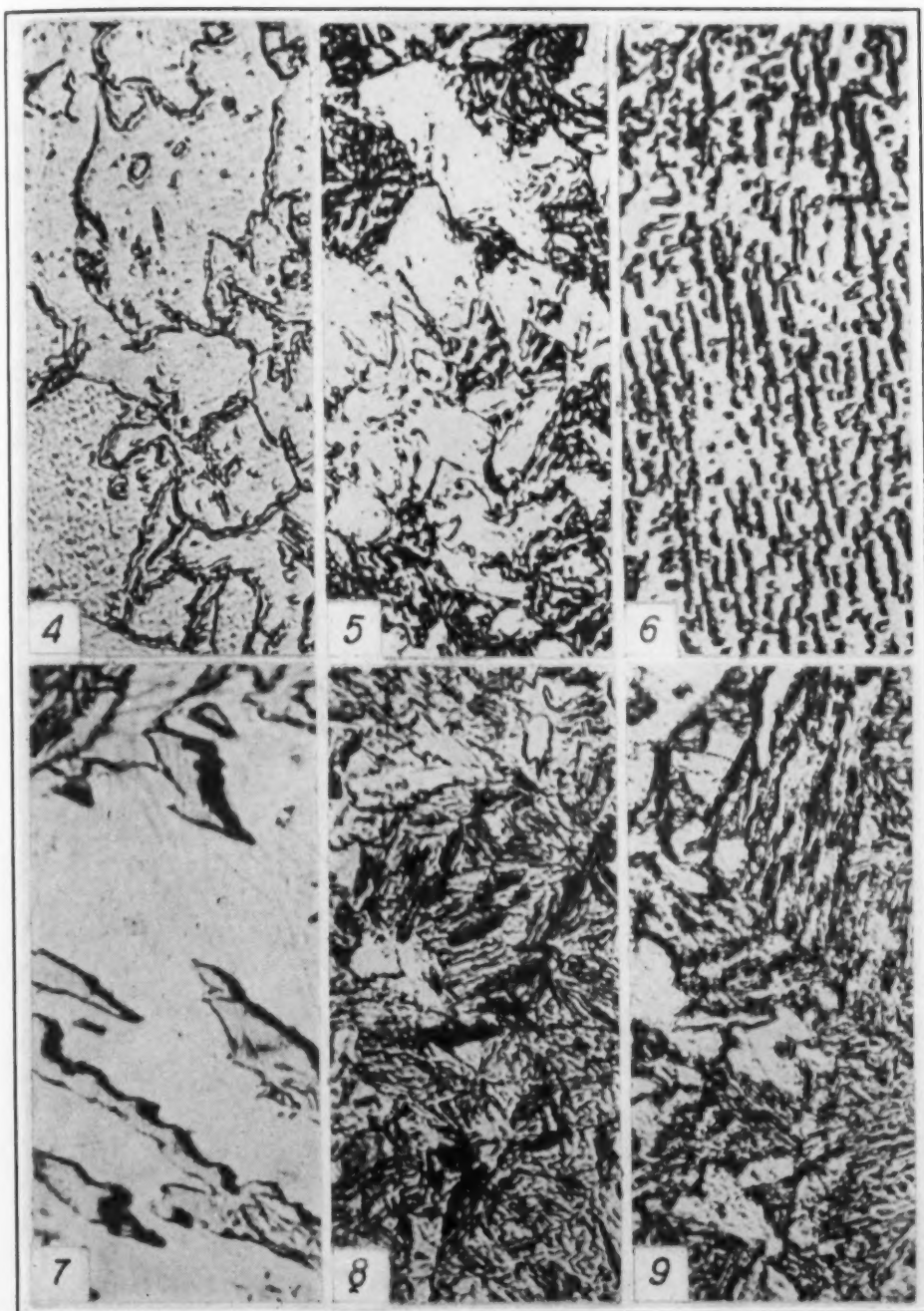Fig. 7—Melt No. 160, Eutectoid 0.07 Per Cent Carbon, 13.48 Per Cent Nickel. As Cast, Annealed. Brinell 229. × 1800.

Fig. 8—Melt No. 136, Hypereutectoid 0.38 Per Cent Carbon, 10.78 Per Cent Nickel. As Cast, Annealed. Brinell 388. × 1000.

Fig. 9—Melt No. 167, Hypoeutectoid, 0.03 Per Cent Carbon, 13.56 Per Cent Nickel. As Cast, Annealed. Brinnell 229. × 1000.

Charges of electrolytic nickel to this steel were made up consisting of 7.9, 8.9, and 9.9 per cent nickel respectively and double-melted in vacuum. The ingots were annealed and examined. A tabular view of the chemical analysis of the commercial nickel steel ingots and the pure nickel steel ingots of similar composition are given in Table III.

Table III
Chemical Composition of Nickel Steel Ingots

| | | Annealed Commercial Nickel Iron-Carbon Ingot after double Melting in Vacuum % Carbon | | | | Annealed Pure Nickel Iron-Carbon Ingot after double Melting in Vacuum % Carbon | | | |
	No.	Com-bined	Graph-itic	% Man-ganese	% Nickel	Com-bined	Graph-itic	% Man-ganese	% Nickel
Eutectoid	578	0.43	none	*0.69	7.9				
Hypereutectoid	577	0.46	none	*0.69	8.77				
Hypoeutectoid	135					0.41	none	*0.079	8.82
Hypereutectoid	576	0.40	none	*0.69	9.53				
Eutectoid	134					0.45	none	*0.079	9.79

*Manganese determined on material as charged. Probably a small amount of manganese was lost in melting.

The results of the microscopic examination of these alloys is summarized as follows:

Melt No. 578 Commercial nickel steel annealed.
0.43 per cent carbon, 0.69 per cent manganese, 7.9 per cent nickel—Eutectoid structure.
Melt No. 577 Commercial nickel steel annealed.
0.46 per cent carbon, 0.69 per cent manganese, 8.77 per cent nickel—Troostito-martensite and sorbite indicating the approach of a hypereutectoid nickel steel (Fig. 12).
Melt No. 135 Pure nickel steel annealed.
0.41 per cent carbon, 0.079 per cent manganese, 8.82 per cent nickel—Ferrite and pearlite decidedly hypoeutectoid (Fig. 4).
Melt No. 576 Commercial nickel steel annealed.
0.40 per cent carbon, 0.69 per cent manganese, 9.53 per cent nickel—One hundred per cent martensitic (hypereutectoid)—(Fig. 13).
Melt No. 134 Pure nickel steel annealed.
0.45 per cent carbon; 0.079 per cent manganese, 9.79 per cent nickel—Eutectoid microstructure (Fig. 6).

Thus a commercial nickel steel containing 0.43 per cent carbon, 0.69 per cent manganese, 7.9 per cent nickel, and a pure nickel steel containing 0.45 per cent carbon, 0.079 per cent manganese, and 9.79 per cent nickel both contain hundred per cent pearlite. In other words 0.61 per cent manganese lowers the nickel in the iron-carbon nickel eutectoid (made of pure iron-carbon bases) about 1.89 per cent.
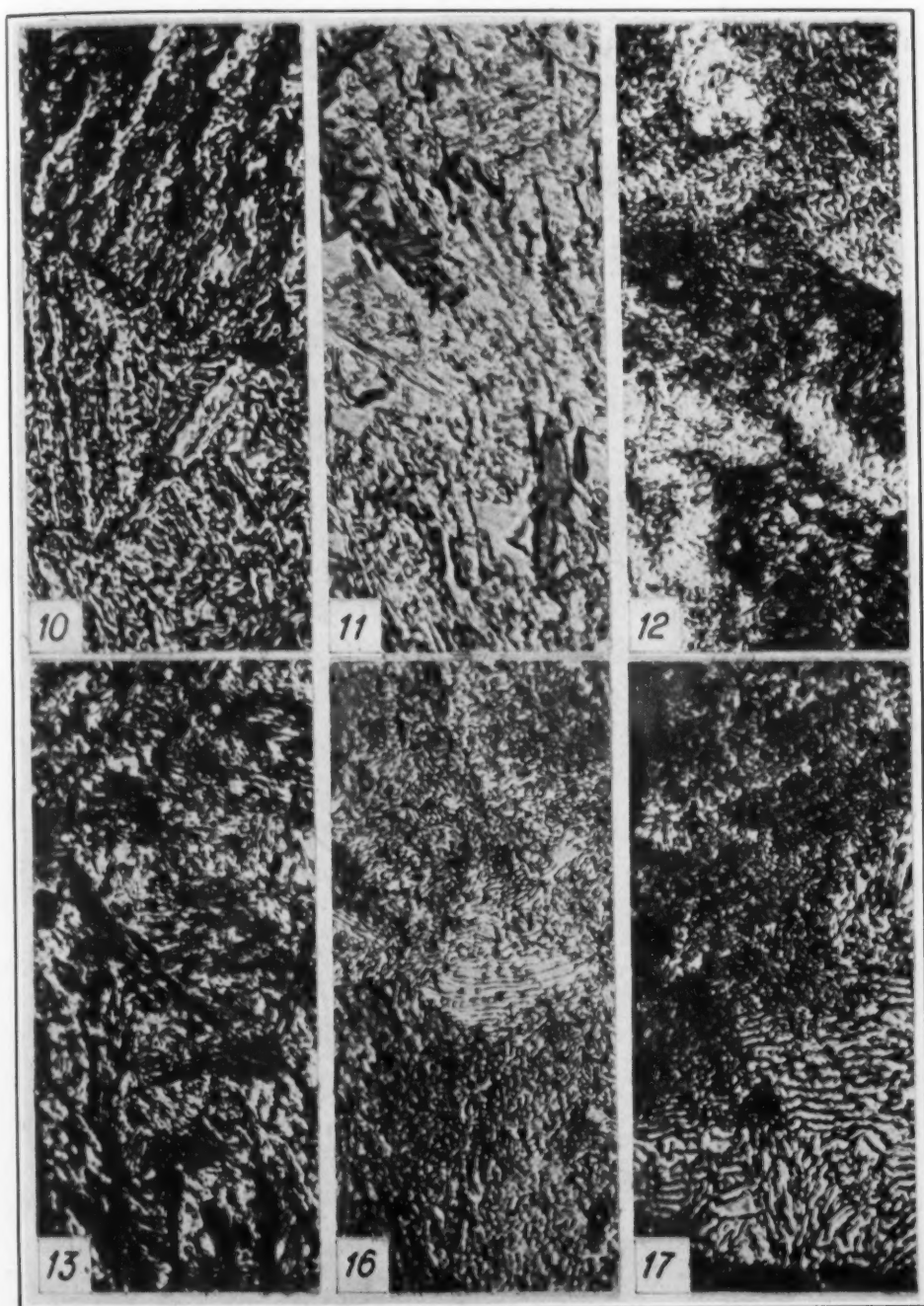
Fig. 10—Melt No. 181, Eutectoid, 0.03 Per Cent Carbon, 14.76 Per Cent Nickel. As Cast, Annealed. Brinell 207. × 1000.

Fig. 11—Air-Melted Annealed Eutectoid Iron-Carbon Nickel Alloy. 0.42 Per Cent Carbon. 9.65 Per Cent Nickel. × 1000.

Fig. 12—Commercial Nickel Steel, Annealed. 0.46 Per Cent Carbon, 0.69 Per Cent Manganese, 8.77 Per Cent Nickel. Troostito-Martensite (Hypereutectoid). Brinell 363. × 1000.

Fig. 13—Commercial Nickel Steel, Annealed. 0.40 Per Cent Carbon, 0.69 Per Cent Manganese, 9.53 Per Cent Nickel. Martensitic (Hypereutectoid). Brinell 461. × 1000.

Fig. 16—Melt No. 544, Hypoeutectoid 0.46 Per Cent Carbon, 2.01 Per Cent Manganese. As Cast, Annealed. Brinell 223. × 1000.

Fig. 17—Melt No. 545, Eutectoid, 0.53 Per Cent Carbon, 2.81 Per Cent Manganese. As Cast, Annealed. Brinell 248. × 1000.

Table IV

	Ingot 537 melted once		Ingot 538 melted twice	
	Carbon Per Cent	Nickel Per Cent	Carbon Per Cent	Nickel Per Cent
Top section annealed	0.49	7.80	0.40	9.87
Middle section annealed	0.50	8.38	0.39	9.93
Bottom section annealed	0.41	17.42	0.39	9.95

Segregation of Carbon and Nickel in Ingots

Double melting the iron-carbon-nickel alloy promotes homogeneity in the melt.

Two melts of iron-carbon-nickel alloys were prepared as follows:

1. Melt No. 537—Charge of 9.9 per cent nickel in Armco base—0.59 per cent carbon melted once.
2. Melt No. 538—Same charge melted twice.

These alloys were annealed and cut into thirds and analyzed for carbon and nickel. The results are shown in Table IV.

Brinell Hardness and Hardening Power of Hypoeutectoid, Eutectoid and Hypereutectoid Iron-Nickel and Iron-Carbon-Nickel Alloys

The hypoeutectoid nickel steels are air hardening steels. They also harden by rapid cooling in the furnace when solidifying from the melt. These hypoeutectoid steels may be substantially annealed (pearlitized) by slow cooling through the critical range.

The eutectoid steels are air hardening steels. They may be made pearlitic or sorbito-pearlitic by cooling slowly through the critical range.

The hypereutectoid steels are martensitic even after slow cooling through the critical range. Omitting the alloys containing 0.03 per cent carbon, maximum hardness after normalizing is exhibited in each series by the alloy of eutectoid composition.

The Brinell hardness of the iron-nickel-carbon steels after annealing and normalizing are shown in Table III. The hardening power values of the nickel steels have also been calculated in Table V.

Thermal Analysis of Hypoeutectoid, Eutectoid and Hypereutectoid Iron-Carbon-Nickel Steels

Critical point specimens were machined from the ingots as pre-

Table V

Brinell Hardness and Hardening Power of Iron-Nickel and Iron-Carbon-Nickel Alloys

No.	Alloy	Chemical Analysis Carbon Per Cent	Nickel Per Cent	As Cast	Annealed	Normalized	Hardening Power
135	hypoeutectoid	0.41	8.82	495	212	461	249
134	eutectoid	0.45	9.79	653	255	601	346
136	hypereutectoid	0.38	10.78	495	388	514	126
158	hypoeutectoid	0.11	10.36	341	262	363	101
145	eutectoid	0.20	11.90	375	269	375	106
144	hypereutectoid	0.22	13.57	363	363	363	0
179	hypoeutectoid	0.10	13.09	302	262	269	7
160	eutectoid	0.07	13.48	277	229	293	65
159	hypereutectoid	0.05	15.03	262	262	285	23
167	hypoeutectoid	0.03	13.56	207	229	241	12
181	eutectoid	0.03	14.76	207	207	217	10
180	hypereutectoid	0.04	15.58	201	229	235	6
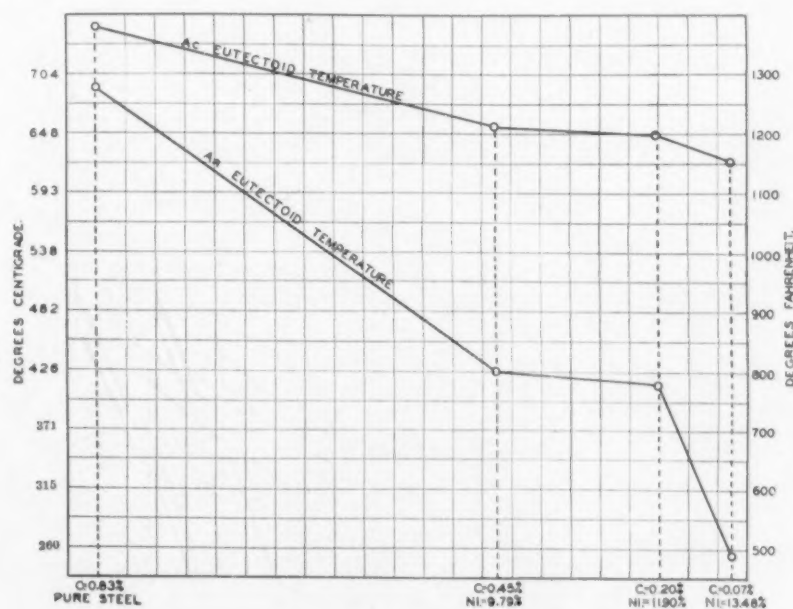
Fig. 14—Nickel-Iron-Carbon Eutectoid Temperatures.

viously described in Fig. 2 and thermal analysis curves drawn by means of a transformation point machine. The results are tabulated in Table VI.

The Ac points are sharper and more pronounced while the Ar points are less indicative of a spontaneous evolution of heat at a definite temperature. Probably the transformation or cooling takes place over a considerable range of temperature.

The cooling curves of the hypereutectoid steels indicate that the Ar″ points are lowered to or below 300 degrees Cent. or the temperature at which the steel becomes martensitic. This fact is in agreement

with the results reported in the microscopic analysis of these alloys. All hypereutectoid nickel iron-carbon alloys are martensitic. Some of the nickel steels have an upper critical point on cooling. The author does not believe these upper points to be cementite points as these specimens were etched in sodium picrate and no indication of free carbide could be detected in the microstructure.

It is believed that these upper points on cooling may be due to the formation of some troostite (the Ar' point). The lower point on
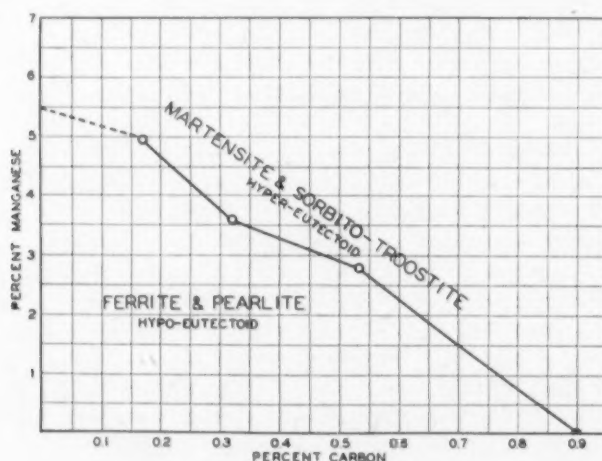
Fig. 15—Influence of Manganese on the Carbon Content of the Iron-Carbon Eutectoid.

cooling is the Ar'' point. With further addition of nickel these upper points should entirely disappear. The author, however, has no results in support of this view.

The effect of nickel on the Ac and Ar temperatures of three

Table VI

Thermal Transformations of Nickel Steels

No.	Alloy	Chemical Analysis Carbon Per Cent	Nickel Per Cent	Ac °C	°F	Ar Upper °C	°F	Lower °C	°F
135	hypoeutectoid	0.41	8.82	653	1210	471	880	471	880
134	eutectoid	0.45	9.79	653	1210			426	800
136	hypereutectoid	0.38	10.78	648	1200			188	370
158	hypoeutectoid	0.11	10.36	664	1230			410	770
145	eutectoid	0.20	11.90	648	1200	681	1260	415	780
144	hypereutectoid	0.22	13.57	631	1170	631	1170	215	420
179	hypoeutectoid	0.10	13.09	659	1220	626	1160	700	371
160	eutectoid	0.07	13.48	620	1150	593	1100	260	500
159	hypereutectoid	0.05	15.03	653	1210	576	1070	349	600
351		0.04	9.79	697	1290			520	950
180		0.04	15.58	642	1190	504	940	315	600
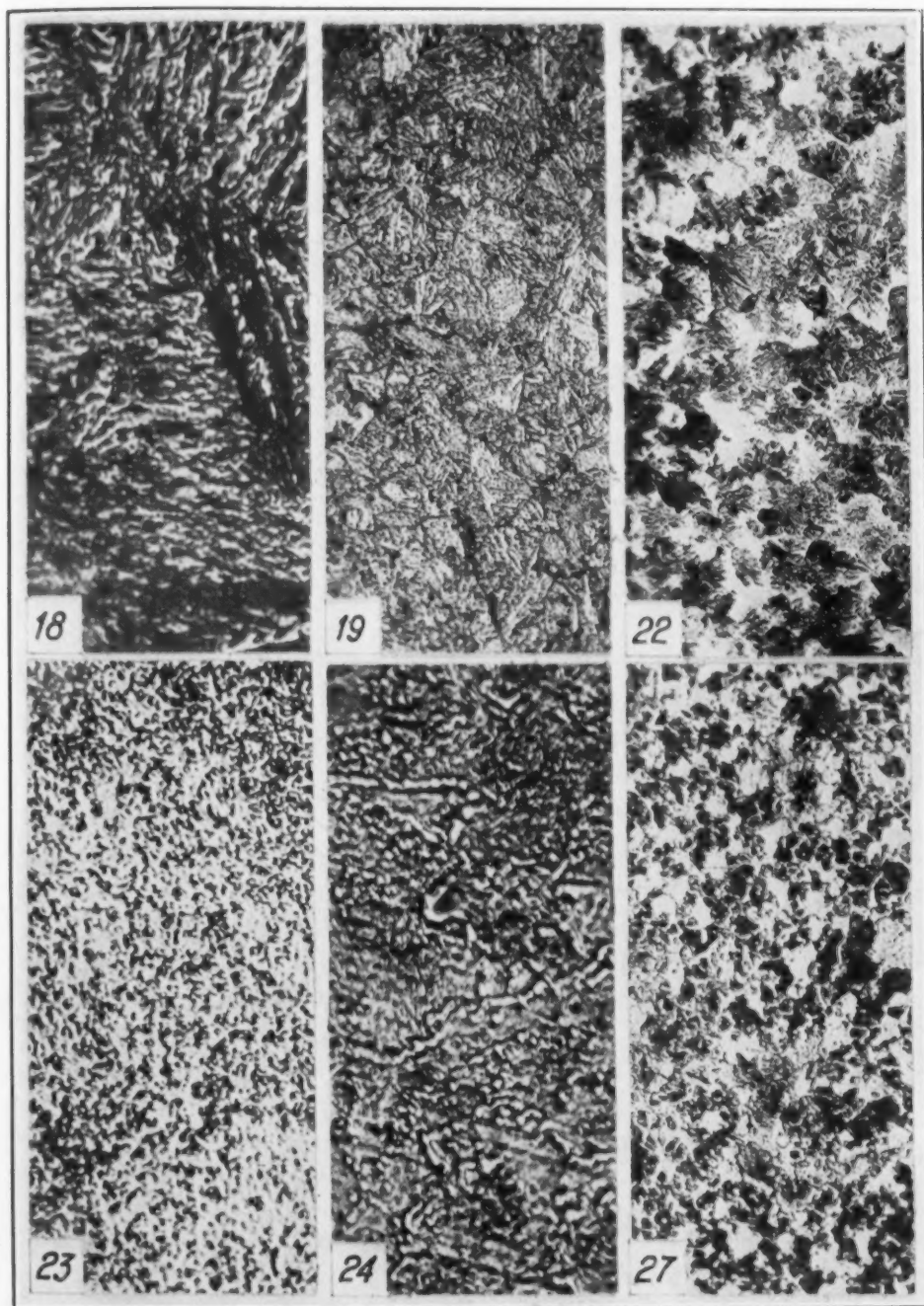
Fig. 18—Melt No. 553, Hypereutectoid, 0.29 Per Cent Carbon, 4.98 Per Cent Manganese. As Cast, Annealed. × 1000.

Fig. 19—Melt No. 557, 0.01 Per Cent Carbon, 4.72 Per Cent Manganese. As Cast, Annealed. Brinell 207. × 100.

Fig. 22—Melt No. 177, Hypoeutectoid, 0.11 Per Cent Carbon, 7.17 Per Cent Chromium. As Cast, Annealed, Brinell 109. × 100.

Fig. 23—Melt No. 418, Eutectoid, 0.28 Per Cent Carbon, 11.73 Per Cent Chromium. As Cast, Annealed, Brinell 153. × 1000.

Fig. 24—Melt No. 561, Hypereutectoid, 0.21 Per Cent Carbon, 13.32 Per Cent Chromium. As Cast, Annealed, Brinell 149. × 1000.

Fig. 27—Melt No. 207, Hypoeutectoid, 0.48 Per Cent Carbon, 6.58 Per Cent Tungsten. As Cast, Annealed, Brinell 179. × 100.

nickel eutectoid steels is graphically illustrated in Fig. 14. The critical temperatures of a pure 0.83 per cent carbon steel are included for reference.

Manganese

Manganese like nickel moves the eutectoid point to the left. The eutectoid displacements determined by microscopic examination and hardening power determinations are plotted in Fig. 15. A comparison of the influence of nickel and manganese on the eutectoid is shown in Table XXXVI. The manganese points were taken from the eutectoid curves.

The annealed hypoeutectoid manganese steels contained ferrite and sorbito-pearlite (Fig. 16). Annealed eutectoid manganese steels were found to be sorbito-pearlitic (Fig. 17).

The annealed medium high carbon-manganese hypereutectoid alloys were sorbito-troostitic while the annealed manganese hypereutectoid alloys containing lower carbon were decidedly martensitic on slow cooling (Fig. 18).

Microstructure of Annealed Iron-Manganese Alloys

A series of annealed iron-manganese alloys were studied in order to determine if one hundred per cent eutectoid consisting of two-phase aggregates persisted between the following manganese contents: 0.01 per cent carbon, 4.72 per cent manganese; 0.01 per cent carbon, 5.44 per cent manganese; 0.01 per cent carbon, 5.81 per cent manganese. Two phases (13) were present in the microstructure as illustrated in Fig. 19. It was difficult to determine a definite eutectoid ratio of the phases within these manganese limits. A complete list of the manganese hypoeutectoid, eutectoid and hypereutectoid alloys investigated is shown in Table VII.

Segregation of Carbon and Manganese in the Ingots

Two melts of iron-carbon-manganese alloys were prepared as follows:

1. Melt No. 548—charge of 4 per cent manganese in Armco base—0.59 per cent carbon melted once.

2. Melt No. 548—charge of 4 per cent manganese in Armco base—0.59 per cent carbon melted twice.

Table VII

Manganese Hypoeutectoid, Eutectoid and Hypereutectoid Alloys Studied

No.	Calculated Charge Armco Iron Carbon Alloys Containing Per Cent Carbon	Manganese Added Per Cent	Alloy Desired	Chemical Analysis Melted Twice in Vacuum and Annealed Carbon Per Cent	Manganese Per Cent
544	0.59	3	Hypoeutectoid	0.46	2.01
545	0.59	4	Eutectoid	0.53	2.81
546	0.59	5	Hypereutectoid	0.49	3.27
547	0.59	6	Hypereutectoid	0.49	3.73
550	0.32	4	Hypoeutectoid	0.31	2.31
551	0.32	5	Eutectoid	0.32	3.60
552	0.32	6	Hypereutectoid	0.34	3.96
553	0.32	7	Hypereutectoid	0.29	4.98
554	0.16	5	Hypoeutectoid	0.17	4.16
555	0.16	6	Eutectoid	0.17	4.95
556	0.16	7	Hypereutectoid	0.17	5.48
557	Armco iron	6	Hypoeutectoid ?	0.01	4.72
558	Armco iron	7	Eutectoid ?	0.01	5.44
559	Armco iron	8	Hypereutectoid ?	0.01	5.81

Chemical analyses of the top, middle and bottom sections of the melts are shown in the following tabulation.

	Ingot No. 548 Melted Once Carbon Per Cent	Manganese Per Cent	Ingot No. 549 Melted Twice Carbon Per Cent	Manganese Per Cent
Top section annealed	0.61	4.73	0.59	3.70
Middle section annealed	0.57	3.46	0.51	3.27
Lower section annealed	0.56	1.27	0.47	2.57

The foregoing experiments substantiate the belief that manganese segregates at the top of the melt. Carbon segregates to a small degree in the upper sections of the ingot, probably as Mn_3C.

Brinell Hardness and Hardening Power of Hypoeutectoid, Eutectoid and Hypereutectoid Iron-Manganese and Iron-Carbon-Manganese Alloys

The following conclusions appear warranted:

1. All the steels are air hardening.
2. The steels of maximum hardening power are manganese eutectoid steels with the exception of the alloy containing 0.17 per cent carbon and 4.95 per cent manganese.
3. The iron-manganese alloys are harder in the annealed state than in the normalized condition.

The decrease in hardness of the normalized iron-manganese alloys may be due to the fact that quick cooling retains the aggregates

composed either of iron-rich or manganese-rich phases partially in solid solution and the resulting hardness of the solid solution may be less than the aggregate of these phases present in the annealed steels.

One may question the soundness of this argument, however, for
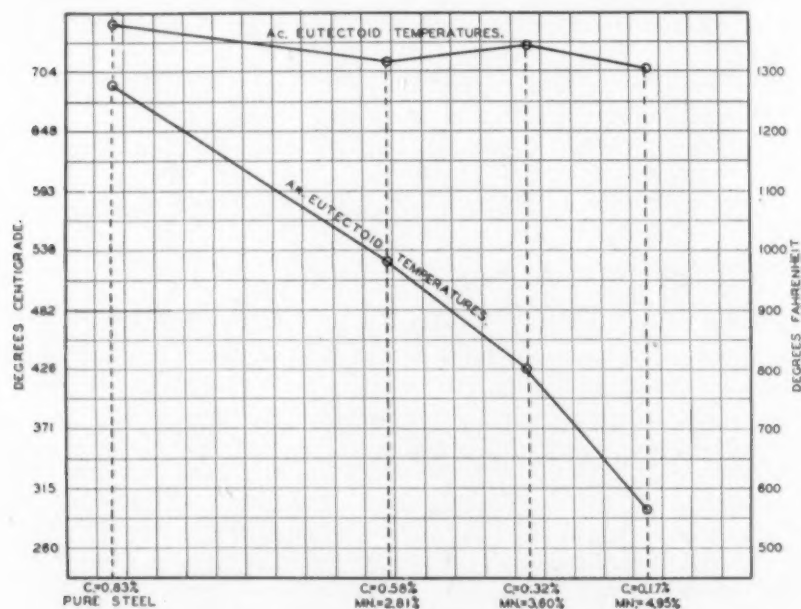
Fig. 20—Manganese-Iron-Carbon Eutectoid Temperatures.

the microstructure of the cast and annealed iron-manganese alloys appear to be composed of aggregates of two phases in about the same proportion. Brinell hardness determinations were made on the ingots "as cast" and after annealing. These results are reported in Table VIII.

Table VIII
Brinell Hardness of Iron-Manganese Alloys

Alloy No.	Chemical Analysis Carbon Per Cent	Manganese Per Cent	As Cast	Annealed	Normalized	Hardening Power
544 Hypoeutectoid	0.46	2.01	321	223	352	129
545 Eutectoid	0.53	2.81	653	248	601	353
546 Hypereutectoid	0.49	3.27	627	255	601	346
550 Hypoeutectoid	0.31	2.31	321	212	341	129
551 Eutectoid	0.32	3.60	514	255	477	222
552 Hypereutectoid	0.34	3.96	461	429	477	48
554 Hypoeutectoid	0.17	4.16	363	269	375	106
555 Eutectoid	0.17	4.95	388	293	388	95
556 Hypereutectoid	0.17	5.48	352	352	363	11
557 Hypoeutectoid?	0.01	4.72	197	207	170	−37
558 Eutectoid?	0.01	5.44	207	217	197	−20
559 Hypereutectoid?	0.01	5.81	212	223	212	−11

Table IX
Thermal Critical Points of Manganese Steels

Alloy No.	Chemical Analysis Carbon Per Cent	Manganese Per Cent	Ac °C.	°F.	Ar Upper °C.	°F.	Lower °C.	°F.
544 Hypoeutectoid	0.35	2.01	734	1350	...	....	604	1120
545 Eutectoid	0.53	2.81	715	1320	...	....	526	980
546 Hypereutectoid	0.49	3.27	719	1330	560	1040	399	750
550 Hypoeutectoid	0.31	2.31	726	1340	...	....	443	830
551 Eutectoid	0.32	3.60	734	1350	...	....	426	800
552 Hypereutectoid	0.34	3.96	719	1330	708	1310	387	730
554 Hypoeutectoid	0.17	4.16	719	1330	...	....	393	740
555 Eutectoid	0.17	4.95	710	1310	...	....	293	560
556 Hypereutectoid	0.17	5.48	704	1300	637	1180	315	600

Thermal Analysis of Hypoeutectoid, Eutectoid and Hypereutectoid
Iron-Carbon-Manganese Steels

The thermal critical points of the manganese steels are shown in Table IX. Manganese, like nickel, lowers the Ar_1 point as shown in Fig. 20. Manganese is similar to nickel in that the Ac_1 point is not lowered to any great extent with increase of manganese.

The hypereutectoid manganese steels, like some of the hypereutectoid nickel steels, undergo a transformation on cooling at some 560 to 637 degrees Cent. (1040 to 1180 degrees Fahr.). The normal Ar transformation occurs at some 315 to 526 degrees Cent. (600 to 980 degrees Fahr.). These upper points in the hypereutectoid manganese steels may be cementite points or indications of troostite transformation (the Ar').

CHROMIUM

Chromium, like nickel and manganese, lowers the percentage of carbon in the eutectoid. The eutectoid displacements determined by microscopic examination and hardening power determinations are plotted in Fig. 21.

Graphical representation of the eutectoid displacement by chromium which is calculated by the author as well as eutectoid curves by Aal (14), Monypenny (15) and Russell (16), are shown in Fig. 21. The relation of the element chromium to nickel and manganese in displacing the carbon content of the eutectoid is given in Table XXXVI.

Hypoeutectoid chromium steels contained ferrite and pearlite (Fig. 22).

Annealed eutectoid chromium steels were sorbito-pearlitic (Fig. 23). The annealed hypereutectoid chromium alloys containing 0.30
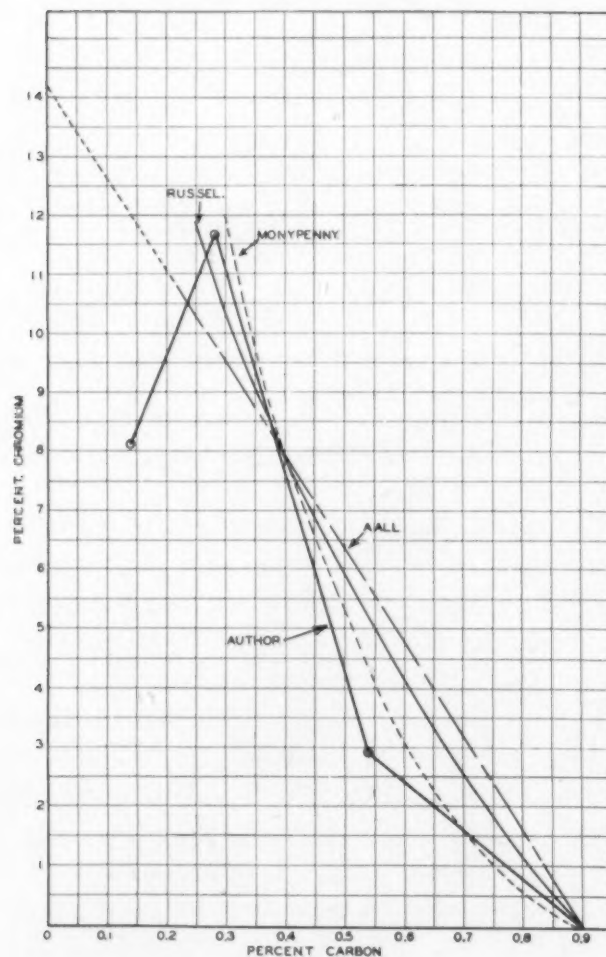
Fig. 21—Influence of Chromium on the Carbon Content of the Iron-Carbon Eutectoid as Determined by Aal, Monypenny, Russel and the Author.

Table X

Chemical Composition of Hypoeutectoid, Eutectoid and Hypereutectoid Chromium Iron Alloys

No.	Calculated Charge Armco Iron-Carbon Alloys Carbon Per Cent	Chromium Added Per Cent	Alloy Desired	Chemical Analysis Melted Once in Vacuum Carbon Per Cent	Chromium Per Cent
174	0.59	2	Hypoeutectoid	0.55	1.90
182	0.59	3	Eutectoid	0.54	2.99
183	0.59	4	Hypereutectoid	0.50	4.01
168	0.32	11	Hypoeutectoid	0.24	11.02
418	0.32	13	Eutectoid	0.28	11.73
561	0.32	13	Hypereutectoid	0.21	13.32
177	0.17	7	Hypoeutectoid	0.11	7.17
171	0.16	8	Eutectoid	0.14	8.13
172	0.16	9	Hypereutectoid	0.10	0.17

per cent carbon bases and over are cementitic and chromium alloys whose iron-carbon bases contain less than some 0.30 per cent carbon and whose chromium content is in excess of 9 per cent, retain alpha (delta) ferrite in the structure (Fig. 24). The chemical analysis of the hypoeutectoid, eutectoid, and hypereutectoid alloys are given in Table X. The loss of carbon and chromium by one melting in vacuum was extremely small.

Segregation of Chromium and Carbon in the Ingots

One melt of an iron-carbon-chromium alloy was prepared as follows:

A charge of 2 per cent chromium in Armco base containing 0.59 per cent carbon was melted once and annealed. The ingot (No. 541
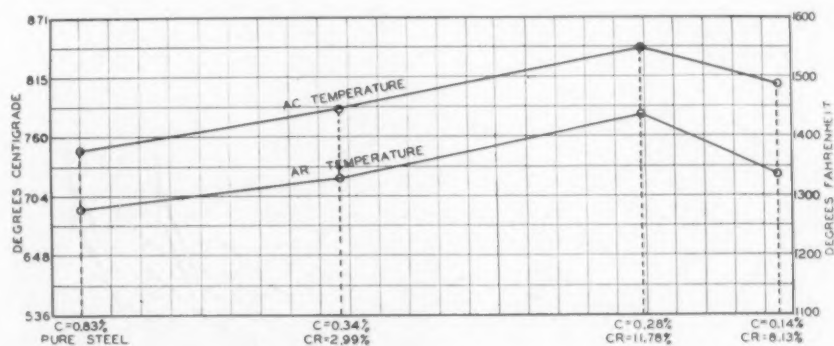
Fig. 25—Chromium-Iron-Carbon Eutectoid Temperatures.

melted once) was cut into thirds and analyzed. The results were as follows:

	Carbon Per Cent	Chromium Per Cent
Top section annealed	0.47	1.91
Middle section annealed	0.47	1.95
Lower section annealed	0.46	1.91

A uniform distribution of carbon and chromium was found in this melt after one melting.

Brinell Hardness and Hardening Power of Hypoeutectoid, Eutectoid and Hypereutectoid Iron-Carbon-Chromium Alloys

Brinell hardness determinations were made on the alloy steels as cast, in the annealed and in the normalized condition (Table XI). The chromium steels were for the most part air hardening. The chromium

Table XI
Chromium Eutectoid Steels

Alloy No.	Chemical Analysis Carbon Per Cent	Chromium Per Cent	As Cast	Annealed	Normalized	Hardening Power
174 Hypoeutectoid	0.55	1.90	321	229	375	146
182 Eutectoid	0.54	2.99	388	241	627	386
183 Hypereutectoid	0.50	4.01	477	241	601	360
168 Hypoeutectoid	0.24	11.02	415	143	388	245
418 Eutectoid	0.28	11.73	555	153	444	291
561 Hypereutectoid	0.21	13.32	388	149	388	138
177 Hypoeutectoid	0.11	7.17	341	109	321	212
171 Eutectoid	0.14	8.13	401	121	363	242
172 Hypereutectoid	0.10	9.17	363	111	331	220
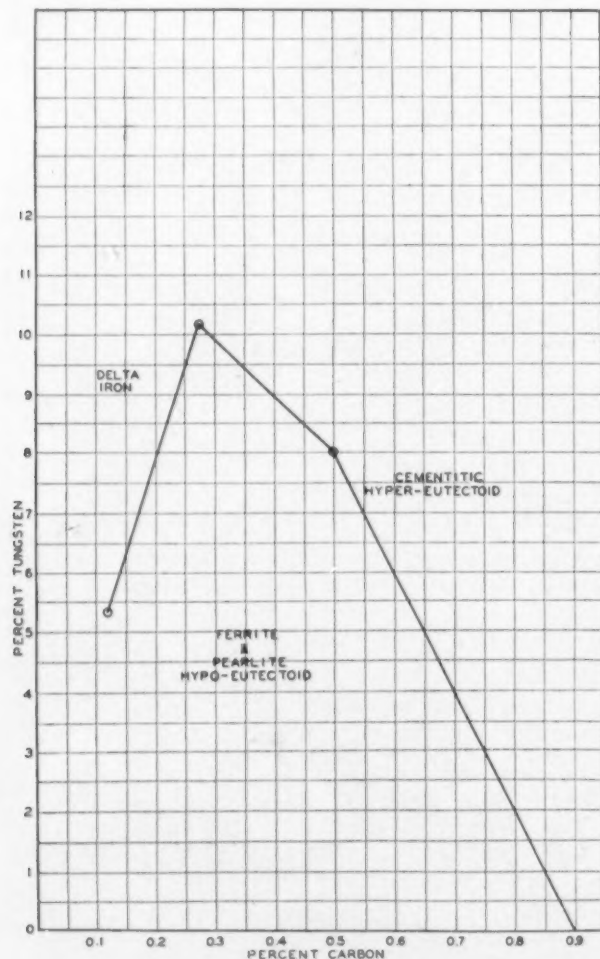
Fig. 26—Influence of Tungsten on the Carbon Content
of the Iron-Carbon Eutectoid.

eutectoid steels possessed maximum hardening power, also greatest
hardness in the normalized condition in each series.

Thermal Analysis of Hypoeutectoid, Eutectoid and Hypereutectoid Iron-Carbon-Chromium Alloys

Critical points of the chromium steels are given in Table XII and the critical temperatures of the chromium eutectoid steels are plotted in Fig. 25. The thermal points in the chromium eutectoid steel containing 0.28 per cent carbon and 11.78 per cent chromium occur at temperatures somewhat above the critical temperatures of the chromium eutectoids containing respectively, 0.54 per cent carbon—2.99 per cent chromium; 0.14 per cent carbon—8.13 per cent chromium. A split or double point was observed in the cooling curves of high chromium steels.

Table XII

Critical Points of Chromium Steels

| | Chemical Analysis | | Ac | | | | Ar | | | |
| | Carbon Per Cent | Chromium Per Cent | Upper | | Lower | | Upper | | Lower | |
Alloy No.	Cent	Cent	°C.	°F.	°C.	°F.	°C.	°F.	°C.	°F.
174 Hypoeutectoid	0.55	1.90	708	1380	776	1430	708	1310	...	....
182 Eutectoid	0.54	2.99	765	1410	787	1450	719	1330	...	....
183 Hypereutectoid	0.50	4.01	765	1510	798	1470	719	1330	...	....
168 Hypoeutectoid	0.24	11.02	737	1360	838	1540	737	1360	...	....
418 Eutectoid	0.28	11.73	737	1360	838	1540	760	1400	708	1310
561 Hypereutectoid	0.21	13.32	726	1340	853	1570	782	1440	704	1300
177 Hypoeutectoid	0.11	7.17	752	1390	815	1500	734	1350	...	....
171 Eutectoid	0.14	8.13	748	1380	809	1440	726	1340	...	....
172 Hypereutectoid	0.10	9.17	741	1370	815	1500	760	1400	734	1350

TUNGSTEN

Tungsten, like chromium, manganese, and nickel, moves the iron-carbon eutectoid point to the left. A graphical representation of the eutectoid displacement by tungsten is shown in Fig. 26. A comparison of the influence of nickel, manganese, chromium and tungsten on the eutectoid displacement is shown in Table XXXVI.

Annealed hypoeutectoid tungsten steels consist of ferrite and sorbite (Fig. 27). In the cast state some of the medium carbon, high tungsten steels were troostito-martensitic. The annealed eutectoid tungsten steels in the annealed condition are sorbitic (Fig. 28).

Annealed hypereutectoid tungsten steels containing high percentages of tungsten and carbon in excess of some 0.30 per cent carbon are cementitic. The delta phase is found in hypereutectoid medium high tungsten low carbon steels (Fig. 29). The chemical analysis of these iron-carbon-tungsten alloys is shown in Table XIII.

Table XIII
Composition of Iron-Carbon-Tungsten Alloys

No.	Calculated Charge Armco Iron-Carbon Alloys Containing Carbon Per Cent	Tungsten Added Per Cent	Alloy Desired	Chemical Analysis Melted Twice in Vacuum and Annealed Carbon Per Cent	Tungsten Per Cent
207	0.59	7	Hypoeutectoid	0.48	6.58
208	0.59	8	Eutectoid	0.48	7.92
209	0.59	9	Hypereutectoid	0.50	8.48
224	0.32	9	Hypoeutectoid	0.25	8.99
222	0.32	10	Eutectoid	0.27	10.17
223	0.32	11	Hypereutectoid	0.25	11.03
230	0.16	5	Hypoeutectoid	0.12	4.91
231	0.16	6	Eutectoid	0.12	5.30
232	0.16	7	Hypereutectoid	0.12	7.43

Segregation of Carbon and Tungsten in the Ingots

Two melts of iron-carbon-tungsten alloys were prepared as follows:

1. Melt No. 539—charge of 8 per cent tungsten in Armco base— 0.59 per cent carbon melted once.
2. Melt No. 540—charge of 8 per cent tungsten in Armco base— 0.59 per cent carbon melted twice.

Chemical analysis of the top, middle and lower sections of the ingots are as follows:

	Ingot No. 539 Melted Once Carbon Per Cent	Tungsten Per Cent	Ingot No. 540 Melted Twice Carbon Per Cent	Tungsten Per Cent
Top section annealed	0.51	1.36	0.53	5.94
Middle section annealed	0.53	7.79	0.43	8.87
Lower section annealed	0.53	13.91	0.43	9.51

It is evident, that carbon does not segregate with tungsten. Tungsten, alone segregates at the bottom of the ingot even after double melting. Possibly a third melting of these alloys would have produced a still more uniform distribution of this element in the ingot.

Brinell Hardness and Hardening Power of Hypoeutectoid, Eutectoid and Hypereutectoid Iron-Carbon-Tungsten Alloys

Brinell hardness determinations were made on the ingots "as cast" and after annealing. These results are reported in Table XIV.

It can be readily inferred from the results in Table XIV that tungsten alone in steel does not promote a marked increase in hardness. To clarify, let us compare the hardness of some normalized
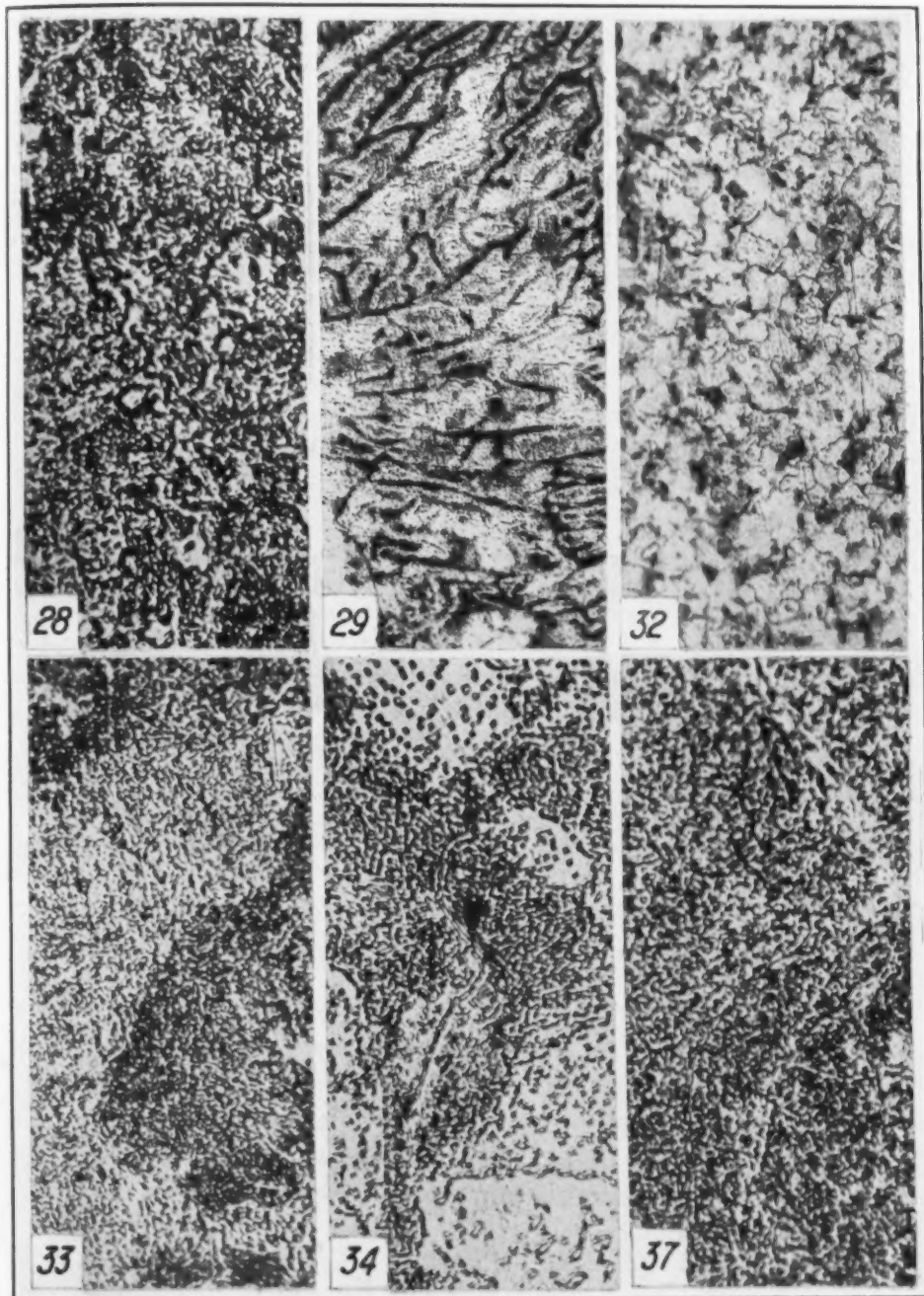
Fig. 28—Melt No. 208, Eutectoid, 0.48 Per Cent Carbon, 7.92 Per Cent Tungsten. As Cast, Annealed, Brinell 197. × 1000.

Fig. 29—Melt No. 232, Hypereutectoid, 0.12 Per Cent Carbon, 7.43 Per Cent Tungsten. As Cast, Annealed, Brinell 143. × 100.

Fig. 32—Melt No. 206, Hypoeutectoid, 0.29 Per Cent Carbon, 5.94 Per Cent Molybdenum. As Cast, Annealed, Brinell 163. × 100.

Fig. 33—Melt No. 205, Eutectoid, 0.33 Per Cent Carbon, 6.54 Per Cent Molybdenum. As Cast, Annealed, Brinell 401, × 1000.

Fig. 34—Melt No. 204, Hypereutectoid, 0.33 Per Cent Carbon, 7.80 Per Cent Molybdenum. As Cast, Annealed, Brinell 156. × 1000.

Fig. 37—Melt No. 308, Hypoeutectoid, 0.32 Per Cent Carbon, 2.27 Per Cent Vanadium. As Cast, Annealed, Brinell 116. × 1000. The Structure of this Alloy Approaches the Eutectoid.

Table XIV
Brinell Hardness Determinations of Iron-Carbon-Tungsten Alloys

No.	Alloy	Chemical Analysis Carbon Per Cent	Chemical Analysis Tungsten Per Cent	As Cast	Annealed	Normalized	Hardening Power
207	Hypoeutectoid	0.48	6.58	285	179	229	50
208	Eutectoid	0.48	7.92	321	197	241	44
209	Hypereutectoid	0.50	8.48	341	197	241	44
224	Hypoeutectoid	0.25	8.99	311	170	179	9
222	Eutectoid	0.27	10.17	311	170	179	9
223	Hypereutectoid	0.25	11.03	321	174	179	5
230	Hypoeutectoid	0.12	4.91	170	143	143	0
231	Eutectoid	0.12	5.30	179	137	137	0
232	Hypereutectoid	0.12	7.43	170	143	149	6

alloy eutectoids which we have already considered: a summary of these results are given in Table XV.

No hardening power values of tungsten eutectoids were obtained.

Table XV
Hardness of Several Normalized Alloy Eutectoids

No.	Alloy	Chemical Analysis in Per Cent	Brinell Hardness Normalized Condition
134	Eutectoid	0.45 Carbon 9.79 Nickel	601
182	Eutectoid	0.54 Carbon 2.99 Chromium	627
545	Eutectoid	0.53 Carbon 2.81 Manganese	601
208	Eutectoid	0.48 Carbon 7.92 Tungsten	241

Thermal Analysis of Hypoeutectoid, Eutectoid, and Hypereutectoid Iron-Carbon-Tungsten Steels

Thermal critical points of tungsten steels are tabulated in Table XVI. The thermal critical temperatures of the iron-carbon-tungsten alloys have been plotted in Fig. 30. Tungsten increases hysteresis in
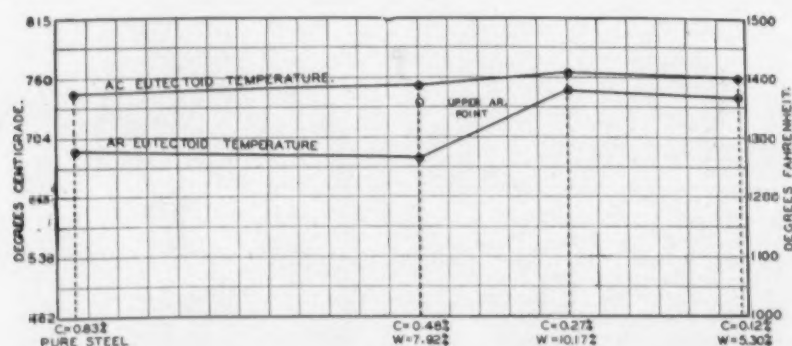

Fig. 30—Tungsten-Iron-Carbon Eutectoid Temperatures.

the alloy steel containing 0.48 per cent carbon and 7.92 per cent tungsten and decreases hysteresis in the alloy eutectoids containing respectively 0.27 per cent carbon, 10.17 per cent tungsten; and 0.12 per cent carbon, 5.30 per cent tungsten. Tungsten slightly raises the Ac point; lowers the Ar point in medium high carbon steels; raises the Ar point in low carbon steels.

Table XVI
Thermal Critical Points of Tungsten Steels

		Chemical Analysis		Ac Lower		Ac Upper		Ar Upper		Ar Lower	
No.	Alloy	% C	% W	°C	°F	°C	°F	°C	°F	°C	°F
207	Hypoeutectoid	0.46	2.01	760	1400			734	1350	686	1270
208	Eutectoid	0.53	2.81	752	1390			734	1350	681	1260
209	Hypereutectoid	0.49	3.27	760	1400			748	1380	681	1260
224	Hypoeutectoid	0.31	2.31	765	1410	871	1600	842	1550	748	1380
222	Eutectoid	0.32	3.60	765	1410	898	1650	1550	842	741	1370
223	Hypereutectoid	0.34	3.96	765	1410	871	1600	1510	820	748	1380
230	Hypoeutectoid	0.17	4.16	760	1400			741	1370		
231	Eutectoid	0.17	4.95	760	1400			737	1360		
232	Hypereutectoid	0.17	5.48	760	1400			741	1370		
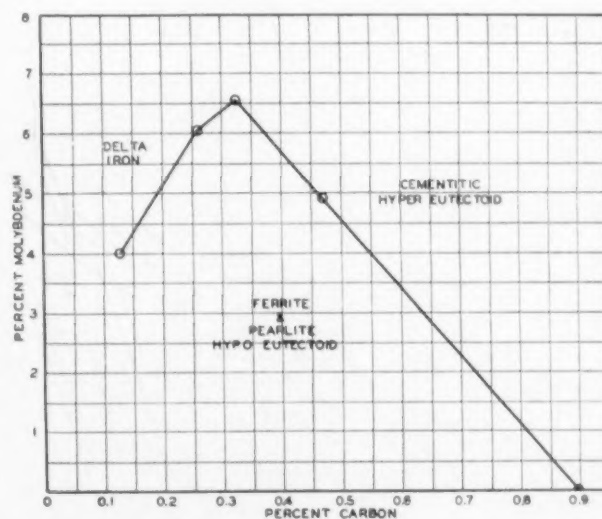
Fig. 31—Influence of Molybdenum on the Carbon Content of the Iron-Carbon Eutectoid.

MOLYBDENUM

Molybdenum, like nickel, manganese, chromium and tungsten, lowers the carbon content of the eutectoid. A graphical representation of the displacement of the iron-carbon eutectoid by molybdenum is shown in Fig. 31. A comparison of the influence of nickel, manganese, chromium and tungsten on the eutectoid displacement is shown in Table XXXVI.

The influence of molybdenum on the eutectoid has been determined by Guillet (17). Guillet finds an eutectoid in a molybdenum steel containing 0.40 per cent carbon and 8 per cent molybdenum. In the present work, however, a molybdenum eutectoid was observed in an alloy containing 0.33 per cent carbon and 6.54 per cent molybdenum. An alloy of 0.33 per cent carbon containing 7.80 per cent molybdenum contained *free delta ferrite* (Fig. 34).

The annealed hypoeutectoid steels contained free ferrite and sorbite (Fig. 32). Molybdenum eutectoid steels are sorbitic in the annealed state (Fig. 33), while annealed hypereutectoid molybdenum-iron-carbon alloys which contain carbon in excess of 0.325 per cent are cementitic.

Alpha (delta) iron is present in annealed hypereutectoid molybdenum steels containing carbon below 0.30 per cent (Fig. 34).

The chemical analysis of the molybdenum alloys investigated is shown in Table XVII.

Table XVII
Chemical Analysis of the Molybdenum Alloys Studied

No.	Calculated Charge Armco Iron-Carbon Alloys Containing Carbon Per Cent	Molybdenum Added Per Cent	Alloy Desired	Chemical Analysis Melted Twice in Vacuum and Annealed Carbon Per Cent	Molybdenum Per Cent
99	0.59	4	Hypoeutectoid	0.47	3.90
98	0.59	5	Eutectoid	0.47	4.95
97	0.59	6	Hypereutectoid	0.45	6.00
206	0.40	6	Hypoeutectoid	0.29	5.94
205	0.40	7	Eutectoid	0.33	6.54
204	0.40	8	Hypereutectoid	0.33	7.80
213	0.32	5	Hypoeutectoid	0.28	5.82
212	0.32	6	Eutectoid	0.26	5.99
211	0.32	7	Hypereutectoid	0.26	9.06
216	0.16	3	Hypoeutectoid	0.11	3.09
215	0.16	4	Eutectoid	0.13	4.07
214	0.16	5	Hypereutectoid	0.13	5.05

Segregation of Carbon and Molybdenum in the Ingots

Two melts of iron-carbon-molybdenum alloys were prepared as follows:

1. Melt No. 542—charge of 4 per cent molybdenum in Armco base —0.59 per cent carbon melted once.

2. Melt No. 543—same charge melted twice.

Chemical analyses determined on ingots No. 542 and No. 543 are as follows:

Table XVIII
Brinell Hardness Determination on Iron-Carbon-Molybdenum Alloys

No.	Alloy	Chemical Analysis Carbon Per Cent	Molybdenum Per Cent	As Cast	Annealed	Normalized	Hardening Power
99	Hypoeutectoid	0.47	3.90	363	223	341	118
98	Eutectoid	0.47	4.95	514	197	341	141
97	Hypereutectoid	0.45	6.00	495	179	302	123
206	Hypoeutectoid	0.29	5.94	285	163	269	106
205	Eutectoid	0.33	6.54	401	163	255	92
204	Hypereutectoid	0.33	7.80	255	156	207	51
213	Hypoeutectoid	0.28	5.82	197	156	163	7
212	Eutectoid	0.26	5.99	197	140	179	39
211	Hypereutectoid	0.26	9.06	269	149	156	7
216	Hypoeutectoid	0.11	3.09	179	116	163	47
215	Eutectoid	0.13	4.07	179	111	167	56
214	Hypereutectoid	0.13	5.05	156	121	143	22

	Ingot No. 542 Melted Once Carbon Per Cent	Molybdenum Per Cent	Ingot No. 543 Melted Twice Carbon Per Cent	Molybdenum Per Cent
Top section annealed	0.49	2.09	0.42	2.04
Middle section	0.52	4.56	0.49	4.52
Lower section	0.52	6.12	0.50	6.07

It is evident from these results that double melting does not promote ideal homogeneity in molybdenum steels. Carbon failed to
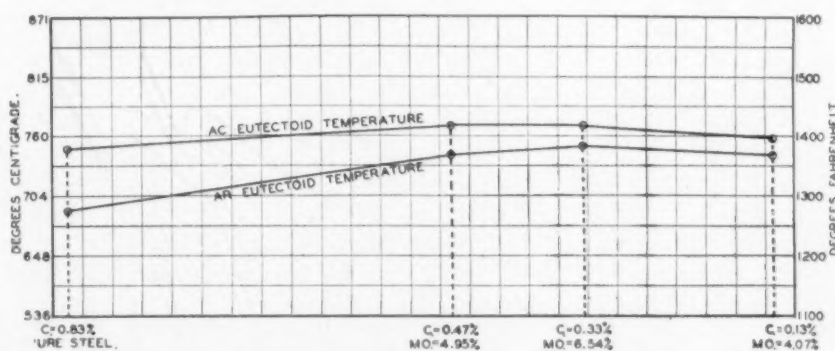
Fig. 35—Molybdenum-Iron-Carbon Eutectoid Temperatures.

segregate appreciably at the bottom of the double melted ingot (No. 543) but molybdenum segregates markedly.

Brinell Hardness and Hardening Power of Hypoeutectoid, Eutectoid, and Hypereutectoid Iron-Carbon-Molybdenum Alloys

Brinell hardness determinations were made on the ingots "as cast" and after annealing. These results are reported in Table XVIII. The maximum hardening power determinations were found in the molybdenum eutectoid steels with the exception of the alloy containing 0.33 per cent carbon and 6.54 per cent molybdenum.

Table XIX
Thermal Analysis Data of Iron-Carbon-Molybdenum Steels

		Chemical Analysis		Ac Lower		Ac Upper		Ar Upper		Ar Lower	
No.	Alloy	% C	% Mo	°C	°F	°C	°F	°C	°F	°C	°F
99	Hypoeutectoid	0.47	3.90	771	1420			760	1400	741	1370
98	Eutectoid	0.47	4.95	771	1420			766	1430	741	1370
97	Hypereutectoid	0.47	6.00	748	1380	827	1520	798	1470	737	1360
206	Hypoeutectoid	0.29	5.94	765	1410			741	1370		
205	Eutectoid	0.33	6.54	771	1420	860	1580	842	1550	737	1360
204	Hypereutectoid	0.33	7.80	752	1390			737	1360		
214	Hypoeutectoid	0.11	3.09	760	1400	882	1620	741	1370		
215	Eutectoid	0.13	4.07	760	1400	831	1530	741	1370		
214	Hypereutectoid	0.13	5.05	752	1390			737	1360		
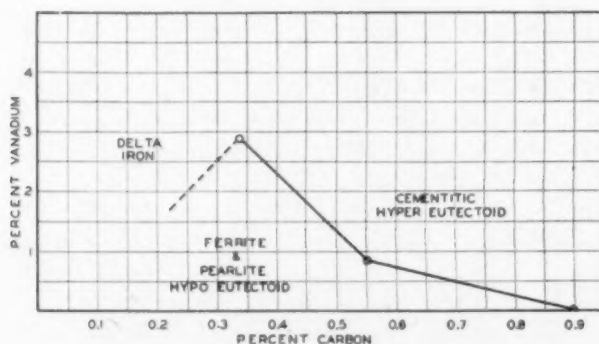
Fig. 36—Influence of Vanadium on the Carbon Content of the Iron-Carbon Eutectoid.

Thermal Analysis of Hypoeutectoid, Eutectoid, and Hypereutectoid Iron-Carbon-Molybdenum Steels

Thermal analysis data is tabulated in Table XIX. The eutectoid temperatures (Ac and Ar) are plotted in Fig. 35. Molybdenum raises the critical range only to a slight degree. Hysteresis is decreased in low and medium carbon-molybdenum steels (0.13-0.33 per cent carbon, 4-7 per cent molybdenum).

VANADIUM

Vanadium, like nickel, manganese, chromium, tungsten and molybdenum, lowers the carbon content of the eutectoid. The vanadium eutectoids determined by microscopic analysis are plotted in Fig. 36.

The eutectoid ratio line shown in Fig. 36 clearly indicates the small percentage of vanadium necessary to cause a displacement of the carbon eutectoid. A comparison of the influence of nickel, manganese, chromium, tungsten, molybdenum and vanadium on the eutectoid displacement is shown in Table XXXVI.
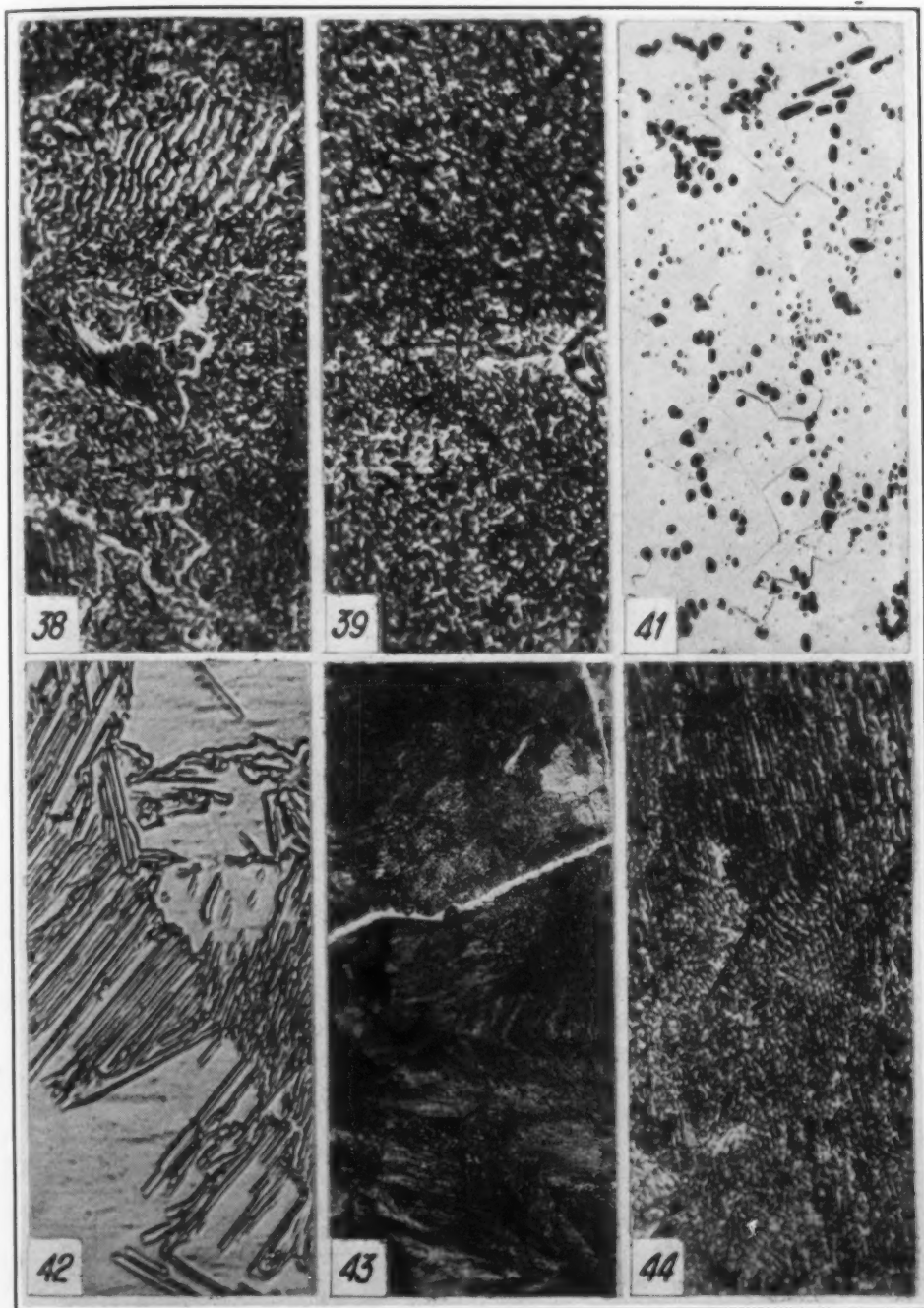
Fig. 38—Melt No. 239, Eutectoid, 0.55 Per Cent Carbon, 1.52 Per Cent Vanadium.   As Cast, Annealed, Brinell 149.   × 1000.   Traces of Free Carbide are Present.

Fig. 39—Melt No. 255, Hypereutectoid, 0.35 Per Cent Carbon, 3.04 Per Cent Vanadium. As Cast, Annealed, Brinell 103.   × 1000.

Fig. 41—Melt No. 303, 0.02 Per Cent Combined Carbon, 0.55 Per Cent Graphitic Carbon, 3.85 Per Cent Silicon (Analysis on Annealed Sample).   As Cast, Annealed, Brinell 197. × 100.

Fig. 42—Melt No. 310, 0.53 Per Cent Total Carbon, 6.76 Per Cent Silicon.   As Cast. × 1000.

Fig. 43—Melt No. 278, 0.52 Per Cent Carbon, 1.94 Per Cent Copper.   Brinell 293. × 100.

Fig. 44—Melt No. 332, Eutectoid, 1.13 Per Cent Carbon, 8.72 Per Cent Copper.   As Cast, Annealed, Brinell 285.   × 1000.

Table XX

Chemical Analysis of Iron-Carbon-Vanadium Alloys Studied

No.	Calculated Charge Armco Iron-Carbon Alloys Containing Carbon Per Cent	Vanadium Added Per Cent	Alloy Desired	Chemical Analysis Melted Twice in Vacuum and Annealed Carbon Per Cent	Vanadium Per Cent
238	0.59	1	Hypoeutectoid	0.55	0.81
239	0.59	2	Eutectoid	0.55	1.52
257	0.59	3	Hypereutectoid	0.49	2.35
308	0.16	3	Hypoeutectoid	0.32	2.27
307	0.16	4	Eutectoid	0.34	2.87
255	0.32	4	Hypereutectoid	0.35	3.03

Annealed hypoeutectoid vanadium steels contain ferrite and sorbito-pearlite (Fig. 37) while annealed eutectoid vanadium steels are sorbito-pearlitic (Fig. 38).

Annealed hypereutectoid medium high carbon-vanadium steels (0.50 per cent carbon—1.50 per cent vanadium) are cementitic and likewise medium high carbon–high vanadium steels (0.50 per cent carbon, 3-4 per cent vanadium) contain free delta ferrite and some carbides (Fig. 39).

The chemical analysis of iron-carbon-vanadium alloys studied is shown in Table XX. There was a considerable loss of vanadium after melting twice in vacuum. The loss of carbon in these vanadium melts was negligible. Some of the low carbon-vanadium charges picked up carbon during melting.

Brinell Hardness and Hardening Power of Hypoeutectoid, Eutectoid, and Hypereutectoid Iron-Carbon-Vanadium Alloys

Brinell hardness determinations were made on the ingots "as cast" and after annealing. (See Table XXI).

Vanadium, like molybdenum and tungsten, does not produce appreciable hardness when present alone in steel. No hardening power

Table XXI

Brinell Hardness Determinations on Iron-Carbon-Vanadium Alloys Studied

No.	Alloy	Chemical Analysis Carbon Per Cent	Vanadium Per Cent	As Cast	Annealed	Normalized	Hardening Power
238	Hypoeutectoid	0.55	0.81	321	163	269	106
239	Eutectoid	0.55	1.52	341	149	255	106
257	Hypereutectoid	0.49	2.35	321	134	255	121
308	Hypoeutectoid	0.32	2.27	149	116	179	63
307	Eutectoid	0.34	2.87	229	107	107	0
255	Hypereutectoid	0.35	3.04	143	103	99.2	—3.8

values of vanadium steels were obtained. There is a decrease in hardness accompanying an increase in delta ferrite in the microstructure of medium and medium high carbon-vanadium steels.

Thermal Analysis of Hypoeutectoid, Eutectoid, and Hypereutectoid Iron-Carbon-Vanadium Steels

Vanadium raises the Ac point slightly. The Ar point is lowered in the vanadium steel containing 0.55 per cent carbon, 1.52 per cent
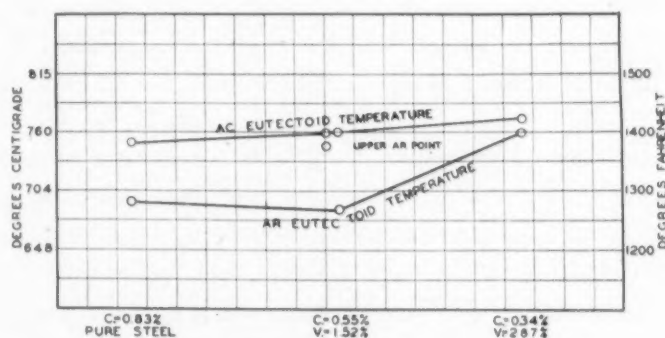
Fig. 40—Vanadium-Iron-Carbon Eutectoid Temperatures.

vanadium and hysteresis is reduced noticeably in the vanadium steel containing 0.34 per cent carbon and 2.87 per cent vanadium.

The critical temperatures of the vanadium steels are plotted in Fig. 40, and the thermal critical points of vanadium steels are given in Table XXII.

Table XXII
Thermal Critical Points of the Vanadium Steels Studied

| | | | | | Ac | | | | Ar | | | |
| | | Chemical Analysis | | Lower | | Upper | | Upper | | Lower | |
No.	Alloy	% C	% V	°C	°F	°C	°F	°C	°F	°C	°F
238	Hypoeutectoid	0.55	0.81	771	1420			741	1370	686	1270
239	Eutectoid	0.55	1.52	760	1400			748	1380	686	1270
257	Hypereutectoid	0.49	2.35	752	1390	871	1600	831	1530	741	1370
308	Hypoeutectoid	0.32	2.27	765	1410	898	1650	853	1570	748	1380
307	Eutectoid	0.34	2.87	771	1420					760	1400
255	Hypereutectoid	0.35	3.04	776	1430					760	1400

SILICON

In the present work it was impossible to determine by microscopic analysis the silicon eutectoid point, either in the cast or annealed silicon steels. *Free ferrite* persisted in the microstructure in the cast

steels or in the annealed steels with constant carbon base of 0.50 per cent carbon and an increase in silicon to a maximum of 6.76 per cent. (Figs. 41 and 42).

In cast steels containing 5.5 to 8.0 per cent silicon, needle shaped crystals of carbide occur, which are possibly a silico-carbide, consisting of cementite and iron silicide. This compound is unstable and readily decomposes into graphite (See Fig. 42). Honda (18), believes that plain cementite does not occur in silicon steels, normally cooled from the melt, which contain more than 5.5 per cent silicon.

In the study of cast silicon steels, the needle shaped carbide was observed within the ranges of silicon given by Honda. Moreover, very small amounts of lamellar pearlite were found in this high silicon range.

The first traces of graphite in the cast state were observed in an alloy containing 0.54 per cent carbon and 2.71 per cent silicon. Graphite was observed in the annealed steel containing 0.56 per cent carbon and 1.85 per cent silicon.

Silicon increases the grain size in medium carbon steels.

The chemical analysis of the silicon steels is given in Table XXIII.

There was neither appreciable loss of carbon nor of silicon in the melt.

Table XXIII
Chemical Analysis of the Silicon Steels Studied

| | Calculated Charge Armco Iron-Carbon Alloy Carbon | Silicon Added | Chemical Analysis Melted Once in Vacuum and Annealed Carbon Per Cent | | | Silicon |
No.	Per Cent	Per Cent	Combined	Graphitic	Total	Per Cent
300	0.59	1	0.52		0.52	0.84
301	0.59	2	0.54	0.02	0.56	1.85
302	0.59	3	0.44	0.10	0.54	2.71
303	0.59	4	0.02	0.55	0.57	3.85
304	0.59	5	0.15	0.38	0.53	4.79
305	0.59	6	0.01	0.45	0.46	5.82
310	0.59	7		0.53	0.53	6.76

Brinell Hardness of Iron-Carbon-Silicon Alloys

Brinell hardness determinations were made on the silicon steels as cast and annealed. The results are shown in Table XXIV.

Table XXIV

Brinell Hardness Determinations of the Silicon Steels Studied

| Alloy No. | Chemical Analysis | | | | | |
| | Carbon Per Cent | | | Silicon | | |
	Combined	Graphitic	Total	Per Cent	As Cast	Annealed
300	0.52		0.52	0.84	241	187
301	0.54	0.02	0.56	1.85	255	217
302	0.44	0.10	0.54	2.71	285	217
303	0.02	0.55	0.57	3.85	302	197
304	0.15	0.38	0.53	4.79	375	269
305	0.01	0.45	0.46	5.82	415	302
310		0.53	0.53	6.76	*100	*100

*Rockwell hardness number "B" scale.

Remarks on the Hardness Tests of Silicon Steels

1.   Silicon progressively increases the hardness of cast medium high carbon steel.

2.   In annealed cast medium high carbon steel, containing from 0.84 to 3.85 per cent silicon, the hardness of these alloys corresponds closely to that of annealed eutectoid carbon steel.

3.   An interesting fact remains, however, that the hardness increases in these same medium high carbon steels with an increase of silicon (from 3.89 to about 6 per cent) although these alloys are nearly completely graphitized.

COPPER

After carefully examining a series of iron-carbon alloys to which copper had been added in increments of 1 per cent (to a maximum of 10 per cent copper) it was found that free ferrite persisted in the microstructure of the cast alloys while in the annealed state, copper promoted a migration of cementite to the grain boundaries.

This phenomenon according to Whiteley (19), is a "marginal coalescence of pearlite". The pearlitic grains which have been depleted of the cementite etched a dark reddish brown in 5 per cent nital. With an increase in copper, there was a greater tendency of the cementite to migrate to the grain boundaries. The microstructure of an iron-carbon-copper alloy containing 0.52 per cent carbon, 1.94 per cent copper as cast and after annealing, is shown in Fig. 43.

The element, copper, according to the results of this investigation failed to displace the eutectoid point to the left. It increases the grain size in cast medium high carbon steels. Therefore several experi-

ments were conducted relative to the possibility of copper moving the eutectoid point to the right.

Copper was added in increments of 1 to 1.4 per cent carbon steel in order to determine if this element would cause the disappearance of cementite. A eutectoid structure was found in the carbon annealed alloy containing 1.13 per cent carbon and 8.72 per cent copper (see Fig. 44). Copper steels containing 1.18 per cent carbon and 7.72 per cent copper, and likewise 1.18 per cent carbon, 8.98 per cent copper, contained free cementite in the cast and annealed state. Small particles of copper-rich iron solution were distributed throughout these high carbon copper steels.

Ishiwara (20), and his associates found a ternary eutectoid in the iron-carbon-copper system, the composition of which was 0.9 per cent carbon and 1.75 per cent copper. This invariant point occurred at 706 degrees Cent. (1303 degrees Fahr.). The structure of the ternary eutectoid, consisted of cementite and a heterogeneous mixture of alpha iron and the particles of copper-rich iron solid solution. The chemical analyses of the copper steels are given in Table XXV.

Table XXV
Chemical Analysis of the Copper Steels Studied

| No. | Calculated Charge | | Chemical Analysis | |
| | Armco Iron-Carbon Alloys Carbon Per Cent | Copper Added Per Cent | Melted Once in Vacuum and Annealed | |
			Carbon Per Cent	Copper Per Cent
279	0.59	1	0.50	1.07
278	0.59	2	0.52	1.94
277	0.59	3	0.49	2.96
276	0.59	4	0.45	3.82
272	0.59	5	0.46	4.66
273	0.59	6	0.43	5.67
274	0.59	7	0.51	6.92
275	0.59	8	0.45	7.67
309	0.59	9	0.45	8.71
283	1.44	8	1.18	7.72
331	1.44	9	1.18	8.98
332	1.44	10	1.13	8.72

Ishiwara and his associates report that steels containing from 1.11 to 1.27 per cent carbon and 3 to 4.85 per cent copper graphitized easily at temperatures exceeding 1100 degrees Cent. (2010 degrees Fahr.). The author failed to observe graphitizing in high copper-high carbon steels which were subjected to the long annealing treatment.

Brinell Hardness of the Iron-Carbon-Copper Alloys

Brinell hardness readings were made on the iron-copper and iron-

Table XXVI

Brinell Hardness Determinations on the Iron-Copper and Iron-Carbon-Copper Alloys Studied

Alloy No.		Chemical Analysis Carbon Per Cent	Copper Per Cent	As Cast	Annealed
279		0.50	1.07	217	152
278		0.52	1.94	293	183
277		0.49	2.96	285	197
276		0.45	3.82	321	207
272		0.46	4.66	388	217
273		0.43	5.67	363	217
274		0.51	6.92	269	183
275		0.45	7.67	341	212
309		0.45	8.71	321	212
283		1.18	7.72	444	277
332	Eutectoid	1.13	8.72	442	285
331		1.18	8.98	415	255

carbon-copper alloys in the cast and in the annealed condition (see Table XXVI).

With an addition of 1 per cent of copper added to a 0.50 per cent carbon cast steel its hardness increases approximately equal to that of an annealed eutectoid steel. In the cast state 4.66 per cent of copper increased the hardness in the same carbon base to a maximum of 388.

Thermal Analysis of Some Iron-Carbon and Iron-Copper Alloys

It has been determined by Clevenger (21), and his co-workers that 4.5 per cent copper lowers the Ar point in carbon steels from 736 to 644 degrees Cent. (1357-1191 degrees Fahr.). Likewise the copper lowers the Ac point in the author's copper-iron-carbon eutectoid (1.13 per cent carbon, 8.72 per cent copper) from 737 to 648 degrees Cent. It appears therefore that copper, after exceeding a certain critical content in steel, fails to lower the Ar point appreciably; that is—

(a) 4.5 per cent copper lowers the eutectoid temperature to 644 degrees Cent.

(b) 8.7 per cent copper lowers the eutectoid temperature to 648 degrees Cent.

Copper lowers the Ac_3 point in iron-copper alloys to some 725 degrees Cent. The Ac_2 point is slightly lowered to this temperature with addition of copper. The Ac_3 and Ac_2 temperatures meet at 725 degrees Cent. corresponding to a composition of nearly 10 per cent copper.

The Ac temperatures are not lowered as copper is increased from 10 to 15 per cent.

The Ar_3 and Ar_2 temperatures meet at some 770 degrees Cent. corresponding to a composition of nearly 5 per cent copper.

The Ar points are not lowered with further increase of copper to a maximum of 15 per cent.

The $Ar_{3.2}$ temperatures occur at some 50 degrees Cent. higher than the $Ac_{3.2}$ transformations.

Aluminum

Aluminum, like copper, fails to move the iron-carbon eutectoid point to the left, in fact the greater percentages of aluminum added to a 0.70 per cent carbon steel the greater proportion of a ferritic constituent in the microstructure. Guillet (22), also observed this phenomenon. For example, the microstructure of an aluminum iron-carbon alloy containing 5.42 per cent aluminum and 0.65 per cent carbon is shown as cast and annealed in Figs. 45, 46. The structure is dendritic while nearly 60 per cent ferrite appears to be present.

Aluminum, like copper, increases the carbon content of the eutectoid. An aluminum-iron-carbon alloy containing 1.31 per cent carbon and 1.30 per cent aluminum had a beautiful lamellar structure in the annealed state (Fig. 47). It was observed that increasing additions of aluminum caused the disappearance of cementite in high carbon steel until the eutectoid structure was determined. Further additions of aluminum-promoted a white constituent resembling ferrite to occur in the cast high carbon steel and after subsequent annealing promoted partial graphitization. A. B. Everest (23), found that on adding between 1 to 8 per cent aluminum to pure white cast iron containing no silicon, a series of soft machinable gray irons is obtained. The microstructure and general physical properties of the castings were found to be analogous to the gray irons in the pure iron-carbon-silicon system.

It was found that 2 to 3 per cent aluminum was required to produce an iron comparable with one containing 1 per cent silicon.

The thermal critical points were determined in the aluminum steel (1.31 per cent carbon, 1.30 per cent aluminum). The Ac transformation occurred at 771 degrees Cent. and the Ar point was detected at 681 degrees Cent. Aluminum raises the critical range.

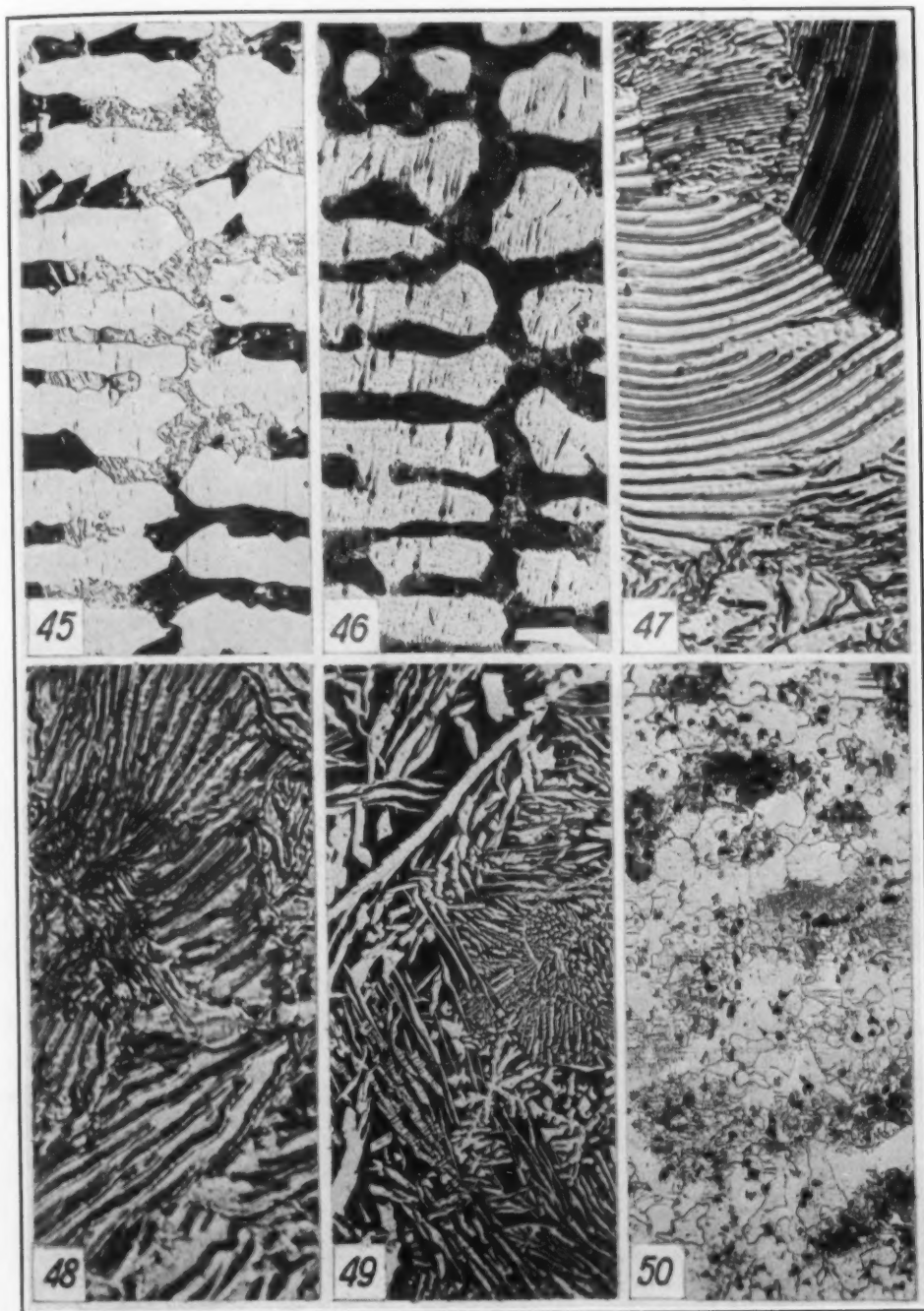Chemical analysis of the aluminum steels is given in Table

Fig. 45—Melt No. 270, 0.65 Per Cent Carbon, 6.42 Per Cent Aluminum, As Cast, Brinell 187. × 100.

Fig. 46—Melt No. 270, 0.65 Per Cent Carbon, 6.42 Per Cent Aluminum. As Cast, Annealed, Brinell 192. × 100.

Fig. 47—Melt No. 336, Eutectoid, 1.31 Per Cent Carbon, 1.30 Per Cent Aluminum. As Cast, Annealed. Brinell 255. × 100.

Fig. 48—Melt No. 325, Eutectoid, 1.24 Per Cent Carbon, 3.03 Per Cent Cobalt (Analysis After Annealing). As Cast, Brinell 255. × 1000.

Fig. 49—Melt No. 315, 0.50 Per Cent Total Carbon. As Cast, Brinell 207. × 100.

Fig. 50—Melt No. 315, 0.34 Per Cent Combined Carbon, 0.16 Per Cent Graphitic Carbon, 0.62 Per Cent Cobalt (Analysis Made on Annealed Sample). As Cast, Annealed, Brinell 131. × 100.

XXVII. With the exception of three high carbon aluminum melts, the chemist reported an increase in carbon after melting twice in vacuum. Chemical analyses of the aluminum steels showed quite a remarkable increase in aluminum in excess of that charged in the melt.

The excess aluminum reported may have been due to segregation of aluminum in portions of the ingot where chemical samples were taken. Aluminum segregated at the top of the ingot to some extent regardless of double melting.

The cause of an increase in carbon in the steel after double melting may be ascribed to the fact that (1) aluminum reacts with oxygen in Armco iron forming alumina; (2) oxygen in the form of alumina cannot oxidize carbon at the melting temperatures of steel; (3) the molten bath may pick up some carbon from the carbon heating coil.

Table XXVII
Chemical Analysis of the Aluminum Steels Studied

No.	Calculated Charge Armco Iron-Carbon Alloys Carbon Per Cent	Aluminum Added Per Cent	Chemical Analysis Melted Twice in Vacuum and Annealed Carbon Per Cent Combined	Graphitic	Aluminum Per Cent
281	0.59	1	0.63	....	1.59
280	0.59	2	0.58	....	2.44
268	0.59	3	0.63	....	2.69
269	0.59	4	0.70	....	5.14
270	0.59	5	0.65	....	6.42
262	0.59	6	0.60	....	6.47
447	1.44	0.5	1.45	....	1.15
336	1.44	1	1.31	....	1.30
335	1.44	2	1.25	....	2.79
334	1.44	3	1.34	0.05	4.06

Brinell Hardness of Iron-Carbon-Aluminum Alloys

Brinell hardness determinations were made on the ingots as cast and in the annealed condition. The results are reported in Table XXVIII.

COBALT

Cobalt, like aluminum and copper, causes a displacement of the eutectoid to the right. Thus, a cobalt steel containing 1.24 per cent carbon and 3.03 per cent cobalt, has a typical eutectoid structure after slow cooling (Fig. 48). Cobalt, in high proportions in medium high and high carbon steels, promotes graphitization. The thermal critical points of this cobalt eutectoid were determined and are Ac —765

Table XXVIII

Brinell Hardness Determinations Made on Iron-Carbon-Aluminum Alloys Studied

| Alloy No. | | Chemical Analysis | | | |
| | | Carbon Per Cent | | Aluminum | |
	Combined	Graphitic	Per Cent	As Cast	Annealed
281	0.63	....	1.59	207	163
280	0.58	....	2.44	223	179
268	0.63	....	2.69	192	179
269	0.70	....	5.14	207	183
270	0.65	....	6.42	187	192
262	0.60	....	6.47	229	212
447	1.45	....	1.15	321	241
336 Eutectoid	1.31	....	1.30	302	255
335	1.25	....	2.79	363	269
334	1.35	0.05	4.06	207	302

degrees Cent. and Ar —710 degrees Cent. The critical range of carbon steels is raised by cobalt.

Cobalt steels are pearlitic. Allison (24) writes that cobalt, like nickel, dissolves in the ferrite and enters into solution with the carbide.

Cobalt graphitizes medium high and high carbon steels. Graphite is detected in annealed steels containing 0.59 per cent carbon and 10 per cent cobalt and also in annealed steel containing 1.01 per cent carbon and 9.06 per cent cobalt. The microstructure of an alloy containing 0.50 per cent carbon and 9.62 per cent cobalt in the cast and annealed condition is shown in Figs. 49, 50.

In high carbon, high cobalt steels, a constituent resembling free

Table XXIX

Chemical Analysis of the Cobalt Steels Studied

| No. | Calculated Charge Armco Iron-Carbon Alloys | | Chemical Analysis Melted Once in Vacuum and Annealed | | |
| | Carbon Per Cent | Cobalt Added Per Cent | Carbon Per Cent | | Cobalt Per Cent |
			Combined	Graphitic	
318	0.59	6	0.49	....	5.82
313	0.59	8	0.50	....	7.99
314	0.59	10	0.44	0.09	8.55
315	0.59	12	0.34	0.16	9.62
316	0.59	14	0.16	0.30	14.20
317	0.59	16	0.22	0.18	16.68
326	1.44	1	1.28	....	1.88
328	1.44	2	1.28	....	1.88
325	1.44	3	1.24	....	3.03
324	1.44	9	1.01	0.12	9.06
323	1.44	11	1.12	0.03	11.04
322	1.44	13	1.05	0.03	13.02
321	1.44	17	1.03	0.02	16.99
445	1.44	22	0.90	0.16	21.96
530	1.44	26	0.31	0.68	26.62
529	1.44	30	0.65	0.31	29.74
514	1.44	40	....	0.75	39.91

ferrite occurred in the microstructure, which was undoubtedly a solid solution of cobalt and iron.

The chemical analyses of the cobalt steels are given in Table XXIX. The loss of cobalt was negligible after one melting in vacuum. With an addition of high cobalt in high carbon steel, a considerable loss of carbon was reported.

Brinell Hardness of Iron-Carbon-Cobalt Alloys

Brinell hardness determinations were made on the ingots as cast, and annealed. (See Table XXX).

Table XXX
Brinell Hardness Determinations Made on the Cobalt Steels Studied

No.	Carbon Per Cent Combined	Graphitic	Cobalt Per Cent	As Cast	Annealed
318	0.49	....	5.82	179	149
313	0.50	....	7.99	217	152
314	0.44	0.09	8.55	197	159
315	0.34	0.16	9.62	207	131
316	0.16	0.30	14.20	170	137
317	0.22	0.18	16.68	197	121
326	1.28	....	1.00	255	207
328	1.28	....	1.88	269	223
325	1.24	....	3.03	255	197
324	1.01	0.12	9.06	331	241
323	1.12	0.03	11.04	302	241
322	1.05	0.03	13.02	321	255
321	1.03	0.02	16.99	341	229
445	0.09	0.16	21.96	341	111
530	0.31	0.68	26.62	363	128
529	0,65	0.31	29.74	302	187
514	....	0.75	39.91	321	166 (1500 Kg)

Melt No. 514 was very brittle in the annealed condition.

TITANIUM AND URANIUM

It appears from the results of this investigation that neither titanium nor uranium have a marked influence on moving the eutectoid to the left. A peculiar grouping of carbides (resembling a eutectic) persist in the free ferrite of the titanium steels (Fig. 51).

Tamaru (25) has drawn a speculative eutectoid line in Fe-Ti-C diagram in which he shows that titanium increases the carbon content of the eutectoid, i.e., an approximate carbon content of 1.1 per cent requires 2.10 per cent titanium to make a eutectoid structure. According to the present author's work, it was determined that a eutectoid is formed which contains 1.27 per cent carbon and 2.10 per cent titanium (Fig. 52).

Free carbide is present in the ferrite areas in cast and annealed uranium steel (Fig. 53). Chemical analyses of the special steels are given in Table XXXI.

Table XXXI
Chemical Analysis of the Titanium and Uranium Steels Studied

| | Calculated Charge | | | Chemical Analysis Melted in Vacuum and Annealed | | |
| | Carbon | Titanium | Uranium | Carbon | Titanium | Uranium |
Alloy No.	Per Cent	Per Cent	Per Cent	Per Cent	Per Cent	Per Cent
564	0.59	2	...	0.38	1.80	....
566	0.59	3	...	0.48	2.17	....
565	0.59	4	...	1.27	2.10	....
527	0.59	.	4.5	0.48	....	1.30
531	0.59	.	4.0	0.56	....	1.77
528	0.59	.	3.5	0.33	....	1.89
568	0.59	.	4.0	0.69	....	2.36
513	0.59	.	...	0.74	....	....
312	0.59	..	...	0.54	....	....
311	0.59	.	...	0.53	....	....
523	0.59	.	...	0.28	....	....

The loss of special element was considerable during the melt. The loss of carbon was noticeable in some melts while in others a marked increase in this element occurred.

Brinell Hardness of the Cast and Annealed Titanium and Uranium Steels

Brinell hardness tests were made on the cast and annealed ingots (see Table XXXII).

Table XXXII
Brinell Hardness Determinations of Titanium and Uranium Steels Studied

Alloy No.	Carbon Per Cent	Titanium Per Cent	Uranium Per Cent	As Cast	Annealed
564	0.38	1.80	....	116	99.2
566	0.48	2.17	....	79.6	88.2
565	1.27	2.10	....	95 Rockwell (B)	248
527	0.48	....	1.30	192	134
531	0.56	....	1.77	197	143
528	0.33	....	1.89	192	143
568	0.69	....	2.36	152	149
513	0.74	....	2.37	241	170
312	0.54	....	....	207	146
311	0.53	....	....	212	152
523	0.28	....	....	170	143

The average hardness of the titanium medium carbon steels is analogous to that of Armco iron.

The element uranium has not increased the hardness of the annealed medium carbon base.
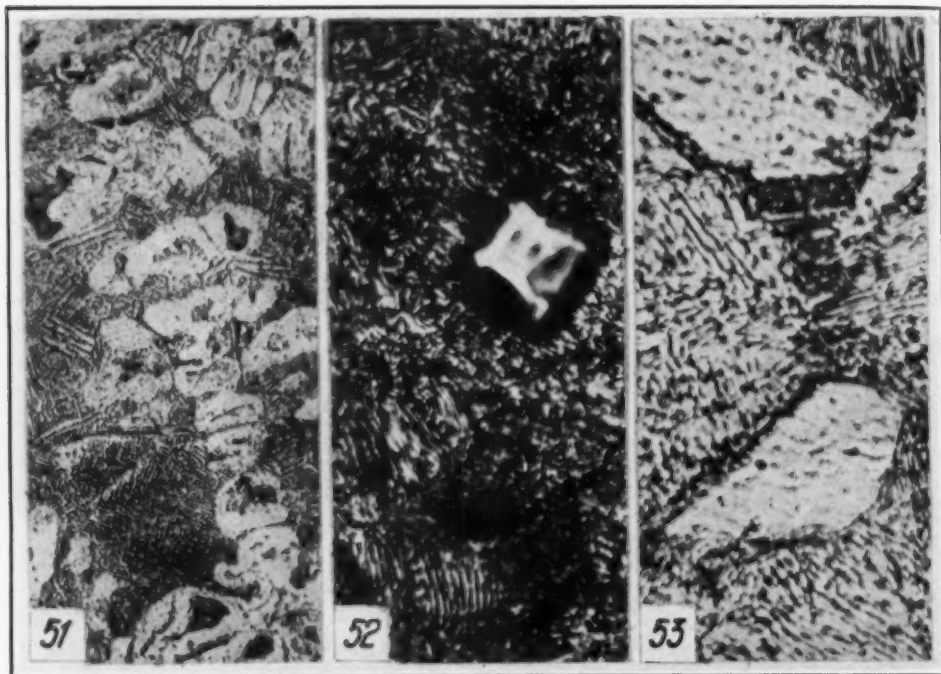
Fig. 51—Melt No. 566, 0.48 Per Cent Carbon, 2.17 Per Cent Titanium, As Cast, Brinell 79.6. × 100.

Fig. 52—Melt No. 565, 1.27 Per Cent Carbon, 2.10 Per Cent Titanium, As Cast, Rockwell 95 B. × 100. The Square Particle in the Field is Probably Titanium Cyananitride.

Fig. 53—Melt No. 568, 0.69 Per Cent Carbon, 2.36 Per Cent Uranium. As Cast, Brinell 152. × 1000.

SPHEROIDIZING OF SOME ALLOY EUTECTOIDS

Some of the alloy eutectoid steels were subjected to the standard spheroidizing treatment and the readiness with which these special steels responded is described.

Preliminary Treatment. Specimens of the alloy eutectoids were heated to 900 degrees Cent. (1650 degrees Fahr.) for ten minutes and quenched in oil. It is believed that spheroidizing takes place more readily if the steel has been previously sorbitized or made martensitic.

Table XXXIII gives the analysis of the eutectoids, the temperature of spheroidizing and the Brinell hardness of the spheroidized steels.

DISCUSSION OF THE SPHEROIDIZED ALLOY EUTECTOIDS

Microstructure. There was practically no difference between the spheroidized structures of the pure carbon eutectoid and the manganese, chromium, tungsten, molybdenum, vanadium, copper and the

Table XXXIII

Analysis of Eutectoids, Spheroidization Temperature and Hardness of Steels Studied

| Alloy No. | Chemical Analysis | | Ac Critical Temperature | Spheroidizing Temperature 7 Hours at | Brinell Hardness Annealed | Brinell Hardness Spheroidized |
	Carbon Per Cent	Special Element Per Cent				
134	0.45	9.79 Ni	1215°F. 657°C.	1132°F. 611°C.	225	461
545	0.53	2.81 Mn	1320°F. 715°C.	1237°F. 669°C.	248	212
182	0.54	2.99 Cr	1450°F. 787°C.	1367°F. 742°C.	241	160.5
208	0.48	7.92 W	1395°F. 757°C.	1312°F. 711°C.	197	207
98	0.47	4.95 Mo	1420°F. 771°C.	1337°F. 725°C.	197	235
239	0.55	1.52 V	1405°F. 762.8°C.	1322°F. 717°C.	149	121
332	1.13	8.72 Cu	1365°F. 740.6°C.	1282°F. 694°C.	285	235
336	1.31	1.30 Al	1430°F. 776°C.	1347°F. 731°C.	255	143
325	1.24	3.03 Co	1400°F. 760°C.	1317°F. 714°C.	197	156
253	0.91 Pure Steel		1375°F. 746°C.	1292°F. 700°C.	207	187

cobalt eutectoids. The carbide particles were dispersed finely throughout the matrix. The nickel eutectoid failed to spheroidize; in other words, the specimen was entirely martensitic subsequent to heating seven hours at 610 degrees Cent. (1130 degrees Fahr.) which was 46 degrees Cent. below the Ac point. The aluminum eutectoid, on the other hand, was completely graphitized.

Brinell Hardness. A decrease in hardness was found in the spheroidized alloy eutectoid steels with the exception of those steels containing molybdenum and tungsten.

PHYSICAL PROPERTIES OF SOME ALLOY STEELS OF APPROXIMATE EUTECTOID COMPOSITION

Nickel, molybdenum and chromium alloy steels with 0.50 per cent carbon bases were very kindly prepared by Dr. M. A. Grossmann. These special steels and a pure carbon eutectoid steel were annealed and subjected to tensile, charpy and twisting tests. The bars submitted for all physical tests were about five-eighths inch square. The analysis of special eutectoid steels is given below.

Heat No.	Carbon Per Cent	Manganese Per Cent	Silicon Per Cent	Chromium Per Cent	Nickel Per Cent	Molybdenum Per Cent
556	0.43	0.70	0.35	....	9.85	....
557	0.52	0.71	0.27	1.91	....	....
558	0.68	0.78	0.23	....	....	4.72

Table XXXIV

Tensile Properties of Special Alloy and Carbon Eutectoid Steels

Heat No.	Chemical Analysis			Tensile Strength (Lbs./Sq.In.)	Yield Point (Lbs./Sq.In.)	Per Cent Elongation in 1.4 In.	Per Cent Reduction in Area	Appearance of Fracture	Heat Treatment
	Carbon Per Cent	Manganese Per Cent	Special Element Per Cent						
556-1	0.43	0.70	9.85 Ni	259,000	170,000	5.7	19.6	Fibrous, cup and cone	Heated 1 hr. at 712.7°C.
556-2				248,000	178,000	5.7	19.6		*Cooled slowly in furnace
557-1	0.52	0.71	1.91 Cr	114,100	67,930	18.6	31.5	Fine, Granular	Heated 1 hr. at 831°C.
557-2				109,100	67,930	18.6	48.0	cup, cone	*Cooled slowly in furnace
558-1	0.68	0.78	4.72 Mo	145,000	90,000	15.0	29.3	Fine grain, cup,	Heated 1 hr. at 827°C.
558-2				123,000	83,000	15.0	38.4	cone	*Cooled slowly in furnace
243-1	0.83% pure steel			109,000	61,200	9.3	15.1	Fine	Heated 25 min. at 910°C.
243-2				106,000	59,100	11.1	15.1	grain	Cooled slowly in *furnace.*
243-1	0.83% pure steel			120,800	71,000	13.6	20.2	Fine	Heated 25 min. at 910°C.
243-2				118,900	71,000	12.9	18.7	granular	Cooled in *air*
243-1	0.83% pure steel			124,700	82,000	20.4	46.3	Fine granular	Heated to 800°C. for 25 min.
243-2				125,500	83,900	18.6	43.0	cup, cone	Quenched in oil, reheated to 650°C. for 10 min., quenched in oil
243-1	0.83% pure steel			105,400	63,500	11.8	12.0	Coarse	Heated to 1000°C. for 1 hr.
243-2				106,400	63,000	9.3	13.5	granular	Cooled very slowly in furnace

*These steels were heated 38 degrees Cent. above the critical point of the pure eutectoids.

The nickel steel (Heat No. 556) contains 0.70 per cent manganese which is sufficient to make this steel a hypereutectoid steel with a martensitic structure. The effect of manganese on the pure iron-carbon-nickel eutectoid was previously studied, where it will be recalled that 0.69 per cent manganese was sufficient to make alloy of 0.40 per cent carbon and 9.53 per cent nickel martensitic. In the absence of manganese, this same alloy would have been pearlitic after slow cooling.

The chromium content of Heat No. 557 was about 1 per cent low. The pure chromium eutectoid contained 0.54 per cent carbon and 2.99 per cent chromium. The loss of chromium may be balanced by the presence of 0.71 per cent of manganese in the steel.

The pure molybdenum eutectoid contained 0.47 per cent carbon and 4.95 per cent molybdenum. The carbon content of Heat No. 3558 is a little high (0.68 per cent). With 0.78 per cent of manganese in this alloy, although the molybdenum content is 4.72 per cent, it is believed this steel is hypereutectoid (cementitic).

Table XXXIV gives the results of tensile tests of the special alloy and pure carbon eutectoid steels and their corresponding heat treatment prior to testing.

SUMMARY OF TENSILE TESTS

1. Steel of maximum strength—nickel steel.
2. Steel of maximum ductility—Double heat treated pure 0.83 per cent carbon steel.
3. Steel possessing a maximum combination of strength and ductility—Double heat treated pure 0.83 per cent carbon steel.
4. Steel possessing a good combination of strength and ductility—molybdenum steel.

Charpy tests made on the heat treated steels are given in Table XXXV.

If the one impact value of the chromium eutectoid is correct, then it is about four times that of an annealed 0.83 per cent carbon pure steel.

The charpy value of the nickel eutectoid steel is exceptionally high for martensitic steel.

TWISTING TESTS

Dr. Ralph W. E. Leiter of the metallurgical department in

Table XXXV

Charpy Impact Tests

Heat No.	Chemical Analysis			Impact Value ft.-lbs. Absorbed	Brinell Hardness	Heat Treatment
	Carbon Per Cent	Manganese Per Cent	Special Element Per Cent			
556-1	0.43	0.70	9.85 Ni	5.32	514	Heated 1 hr. at 712.7°C
556-2				3.97	495	Cooled slowly in furnace
557-1	0.52	0.71	1.91 Cr	9.80	269	Heated 1 hr. at 831°C
557-2				7.73	262	Cooled slowly in furnace
558-1	0.68	0.78	4.72 Mo	7.73	285	Heated 1 hr. at 827°C
558-2				7.73	285	Cooled slowly in furnace
243-1	0.83% C pure steel			2.58	207	Heated 25 min. at 910°C
243-1				2.34	212	Cooled slowly in furnace
243-1	0.83% C pure steel			1.91	241	Heated 25 min. at 910°C
243-1				3.32	241	Cooled in air
243-1	0.83% C pure steel			2.39	241	Heated 25 min. at 800°C
243-1				2.06	241	Quenched in oil, reheated to 650°C, 10 min., quenched in oil
243-1	0.83% C pure steel			2.16	207	Heated 1 hr. at 1000°C
243-1				2.16	207	Cooled very slowly in furnace

Harvard University conducted some twisting tests on the alloy and pure carbon steel eutectoids. The description of the apparatus and the details of the method used may be found in the fourth Edward de Mille Campbell Memorial Lecture, "Steel at Elevated Temperatures," presented by Professor Albert Sauveur before the eleventh annual convention of the American Society for Steel Treating, September, 1929, TRANSACTIONS, vol. 17, 1930, p. 410.

It is necessary, however, to explain some of the data obtained in this test. Quoting from Professor Sauveur's paper:

> "The data obtained include: 1. Torsional strength expressed in pounds; that is, the actual weight to break the bar. 2. the angle of rupture; that is, the number of twists or fraction of a twist. 3. the 'factor of stiffness' obtained by dividing the stress by the strain, i.e. the breaking load by the twist."

The summary of twisting tests is given below:

Graphical representation of the twisting tests is given in Figs. 54, 55, 56 inclusive. To summarize:

1. The molybdenum steel possesses the maximum torsional strength at high temperature.

2. The chromium steel is not strengthened in the blue heat range.

3. The ductility of the chromium steel is maximum at low and at high temperatures.

4. There is a marked increase in the "factor of stiffness" curve of the molybdenum and pure carbon steels in the blue heat range.

5. There is a sudden increase in torsional strength in the nickel and pure carbon steels when tested in the blue heat range.
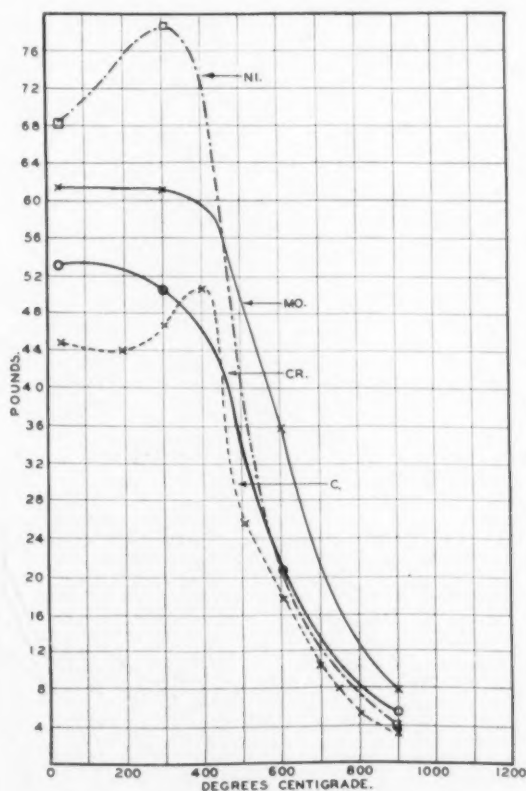

Fig. 54—Torsional Strength of Some Alloy Steels of Approximate Eutectoid Composition at Various Temperatures.

GENERAL DISCUSSION AND CONCLUSION

The Influence of Special Elements on the Carbon Content of the Eutectoid

Since the composition of the alloy eutectoids was determined on vacuum-melted pure iron-carbon bases (containing a low percentage of manganese, phosphorus, sulphur and silicon), we might justly ask what is the role of oxygen on the eutectoid displacement. It will be recalled that an annealed vacuum-melted eutectoid nickel steel containing 0.45 per cent carbon and 9.79 per cent nickel was pearlitic

(Fig. 6) while an annealed air-melted eutectoid nickel steel containing 0.42 per cent carbon and 9.65 per cent nickel was sorbitic (Fig. 11). We might infer from this one experiment that oxygen decreased slightly the amount of nickel necessary to displace the eutectoid point and also oxygen prevented a complete austenito-pearlitic transformation. It was found that a pure 0.99 per cent carbon steel
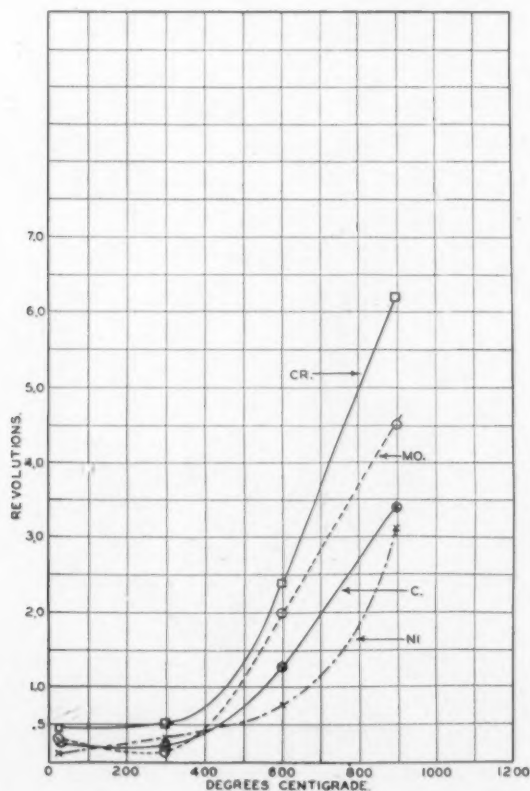
Fig. 55—Angle of Rupture of Some Alloy Steels of Approximate Eutectoid Composition Twisted at Various Temperatures.

melted in vacuum contained 0.003 per cent oxygen and 0.001 per cent nitrogen. Unfortunately, the author has no oxygen analyses on the same material melted in air. On the other hand the oxygen content of Armco iron is decreased by vacuum melting, for example, Armco iron as received contained 0.080 per cent oxygen and 0.003 per cent nitrogen while after vacuum melting it contained 0.028 per cent oxygen and 0.003 per cent nitrogen.

The present work does not include a discussion of the slope of the displaced ferrite and solubility lines. Aal (26) has conducted

some experiments on this phase of the subject and has reported his
results as follows:

1. The displacement of the iron-carbon eutectoid (Point S, Fig.
   57) by a special element does not necessarily mean a change in
   direction of the ferrite and cementite solubility lines.   The
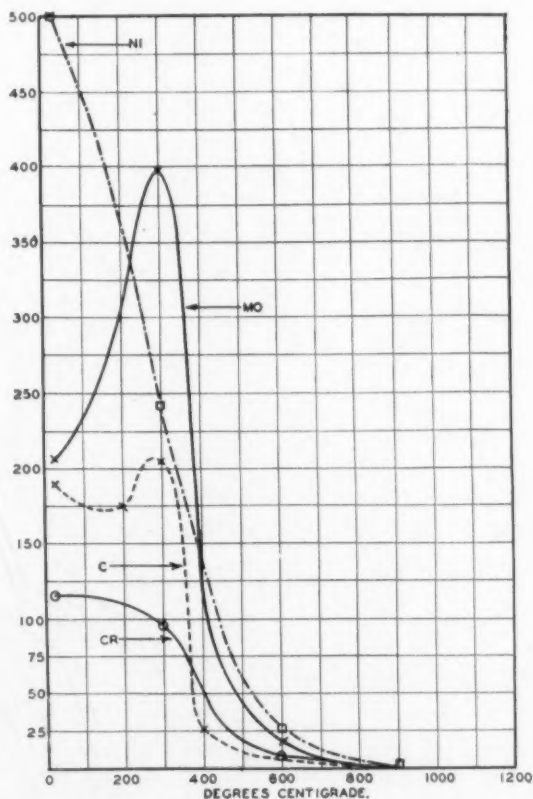   movement of these lines is parallel to their original directions.

Fig. 56—Factor of Stiffness of Some Al-
loy Steels of Approximate Eutectoid Com-
position.

2. Metals forming a solid solution with alpha iron but not form-
   ing carbides move the ferrite and cementite lines parallel to
   their original direction in such a way that the pearlite point
   moves along a straight line SF and crosses the temperature
   axis at absolute zero.

*3. The elements forming carbides but not soluble in alpha iron
   have no influence on the position of the ferrite line but cause

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*Aal does not mention the class of elements which are not soluble in alpha ferrite.

a movement of the cementite line parallel to its original position in such a manner that point S moves along the ferrite line SG.

4. The elements forming carbides and likewise soluble in alpha iron (the carbon percentage being constant) causes the movement of the ferrite and cementite lines (parallel to their former directions) in such a way that the point S (Fig. 57)
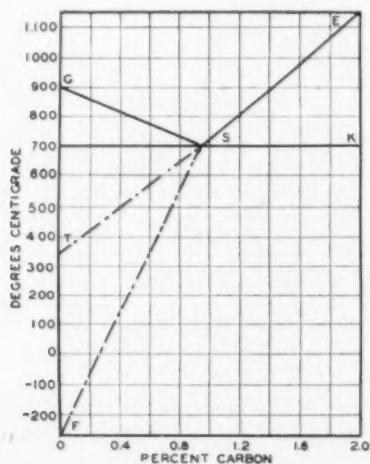

Fig. 57—Equilibrium Diagram of Transformations in Ternary Steels, After Aal.

moves along the straight line ST which crosses the temperature axis at temperature T situated between 900 degrees Cent. and absolute zero. The greater the ability of the alloying element to form carbides, the higher the temperature is raised.

It has been determined that special elements which have a body-centered cubic lattice shift the eutectoid point in the iron-carbon equilibrium diagram to the left, while the face-centered elements (except nickel) displace the eutectoid point to the right. In other words, body-centered elements decrease the eutectoid carbon content and the face-centered elements increase the carbon required to produce a eutectoid structure. The displacement of the eutectoid by some special elements has been graphically constructed in Fig. 58 and also has been tabulated in Table XXXVI.

The author found that the lowering of the carbon content of the eutectoid was not proportional to an increase of alloying element.

EFFECTIVENESS OF SPECIAL ELEMENTS IN DISPLACING THE EUTECTOID POINT

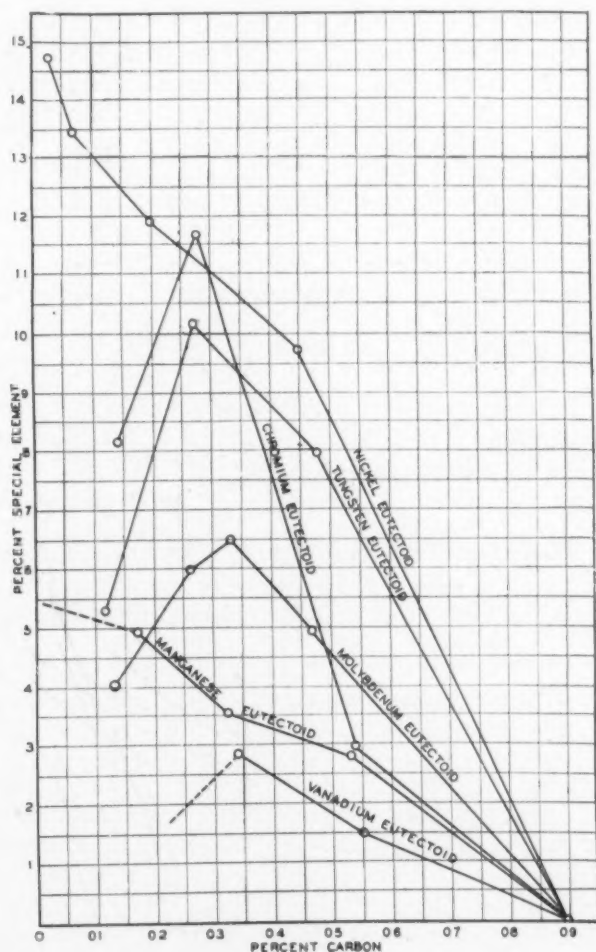The special elements are listed below in order of their effective-

Fig. 58—Influence of Special Elements on the Carbon Content of Iron-Carbon Eutectoid.

ness in displacing the eutectoid point in pure iron-carbon bases containing from 0.40 to 0.80 per cent carbon inclusive:

         1.   Vanadium
         2.   Manganese
         3.   Chromium
         4.   Molybdenum
         5.   Tungsten
         6.   Nickel

Table XXXVI

Comparison of the Influence of Alloying Elements on Eutectoid Displacement

Carbon Per Cent	Nickel Per Cent	Manganese Per Cent	Chromium Per Cent	Tungsten Per Cent	Molybdenum Per Cent	Vanadium Per Cent	Copper Per Cent	Aluminum Per Cent	Cobalt Per Cent
0.1	13.2	5.2	....	....	....	....	....	....	....
0.2	11.95	4.80	9.75	7.80	5.1	....	....	....	....
0.3	11.10	3.75	11.15	9.90	6.35	....	....	....	....
0.4	10.20	3.25	7.65	9.0	5.70	2.25	....	....	....
0.5	8.70	2.90	4.20	8.0	4.55	1.35	....	....	....
0.6	6.50	2.25	2.45	6.0	3.40	0.75	....	....	....
0.7	4.35	1.50	1.60	4.0	2.25	0.50	....	....	....
0.8	2.50	0.75	0.75	2.0	1.10	0.25	....	....	....
0.9	0.00	0.00	0.00	0.00	0.00	0.00	....	....	....
1.13	....	....	....	....	....	....	8.72	....	....
1.24	....	....	....	....	....	....	....	....	3.03
1.31	....	....	....	....	....	....	....	1.30	....

In Table XXXVI we note that it requires 8.7 per cent nickel or 1.35 per cent vanadium to produce a 100 per cent eutectoid structure in a pure 0.50 per cent carbon steel, or in other words, an amount of nickel which is 6.4 times that of vanadium.

ELEMENTS ENLARGING GAMMA FIELD VERSUS DISPLACEMENT OF THE EUTECTOID

In addition to the fact that progressive additions of nickel and manganese lower the percentage of carbon required to produce a true eutectoid structure, it should be remembered that these two elements enlarge the gamma field in iron alloys (with and without carbon). Cobalt, although enlarging the gamma field, appears to increase the carbon content of the eutectoid.

ELEMENTS CLOSING GAMMA FIELD VERSUS DISPLACEMENT OF THE EUTECTOID

In the case of chromium, tungsten, molybdenum and vanadium, the carbon content of the eutectoid is lowered with an increase of special element only to a limited extent. In iron-carbon bases containing less than 0.30 to 0.35 carbon, there is a decided lowering of the special element required to produce a true eutectoid structure. To the left of the sharp drop in the eutectoid curves of chromium, tungsten, molybdenum and vanadium implies the presence of some delta iron. These elements suppress the alpha (delta)-gamma transformation in iron alloys. In alloys containing a low percentage of car-

bon there is a decided increase in special element necessary in order to eliminate the alpha (delta) gamma transformation.

Although aluminum closes the gamma loop, it increases the carbon content of the eutectoid.

Copper, like aluminum, increases the carbon content of the eutectoid but only partially enlarges the gamma field due to its incomplete solubility in gamma iron.

Microstructure Cast Annealed. Some of the annealed nickel, chromium, aluminum and copper eutectoids were composed of lamellar pearlite. The majority of the annealed alloys, however, were sorbitic or sorbito-pearlitic.

To summarize:

1. No free cementite was found in hypereutectoid nickel steel. Hypereutectoid nickel steel is martensitic.
2. Low carbon hypereutectoid manganese steel is martensitic.
3. Hypereutectoid steels containing the carbide forming elements were cementitic.
4. All hypoeutectoid steels contained free ferrite.

CLASSIFICATION OF ELEMENTS WHICH PROMOTE GRAPHITIZING AND WHICH ARE CARBIDE FORMING

Elements Promoting Graphitizing	Elements Carbide Forming
Aluminum	Chromium
Silicon	Tungsten
Nickel	Molybdenum
Cobalt	Vanadium
Copper (according to Ishiwara [27])	Uranium

Homogeneity in Melt by Double Melting. In most cases segregation of the element was eliminated in the ingot by double melting. In others some segregation remained as shown below:

In nickel steel ingots, segregation was nil.

In manganese steel ingots, some segregation persisted at the top.

In tungsten steel ingots some segregation persisted at the bottom.

In molybdenum steel ingots considerable segregation persisted at the bottom.

Brinell Hardness of Cast and Annealed Alloy Eutectoid Steel

1. Nickel, chromium and manganese alloy eutectoid steels were air hardening steels.
2. The Brinell hardness of the annealed alloy eutectoids (with the exception of the copper eutectoid steel) was 255 or less.
3. The hardening power was generally found to be maximum at the eutectoid composition.
4. The hardness of the annealed eutectoid was lowered appreciably by the spheroidizing treatment.

Thermal Analysis

The results of the thermal analysis of alloy steels containing 0.50 per cent and 1.1 per cent carbon bases are given below:

1. Hysteresis is increased by nickel, manganese, copper, aluminum, vanadium and tungsten.
2. Hysteresis is decreased by chromium and cobalt.
3. Nickel and manganese lower the Ac and Ar points.
4. Copper lowers the Ar point without altering the Ac point appreciably.
5. Aluminum, vanadium and tungsten raise the Ac point and lower the Ar point.
6. Molybdenum and cobalt raise the Ac and Ar points. Molybdenum reduces hysteresis to a minimum.

Physical Properties of Some Alloy Steels of Approximate Eutectoid Composition

It was determined that a steel containing 0.68 per cent carbon, 0.78 per cent manganese and 4.72 molybdenum possessed maximum strength at high temperatures. Maximum ductility at high temperatures was found in a steel containing 0.52 per cent carbon, 0.71 per cent manganese and 1.91 per cent chromium. No blue heat phenomenon was observed in this steel. The molybdenum and chromium steels had a good combination of strength and ductility in the annealed state.

Acknowledgments

The author wishes to express his appreciation and thanks to

Professor Albert Sauveur, Gordon McKay Professor of Metallurgy and Metallography at Harvard University, for his interest and helpful suggestions during the progress of the work.

He also wishes to record his sincere thanks for material, apparatus and technical information very courteously given by the following:

Dr. Paul Merica, assistant to the president of the International Nickel Company, New York, who kindly furnished electrolytic nickel for use in making iron-carbon-nickel alloys; Aluminum Company of America, who furnished the aluminum for making the aluminum-iron-carbon alloys; J. H. Critchett, of the Union Carbide Company, New York, who sent the author a large amount of electrolytic chromium for use in making iron-carbon-chromium alloys; W. J. Beck, of the American Rolling Mill Company, Middletown, Ohio, who very kindly made four fifty-pound ingots of pure iron-carbon steels containing 0.16 per cent carbon, 0.32 per cent carbon, 0.59 per cent carbon, and 1.44 per cent carbon; Dr. M. A. Grossmann, then vice president of the Republic Steel Research Corporation, Canton, Ohio, who made three heats of alloy eutectoid steels and rolled them down to five-eighths inch square rods suitable for physical testing; Dr. V. N. Krivobok, associate professor of metallurgy at the Carnegie Institute of Technology, Pittsburgh, who kindly sent the author a good supply of distilled manganese of high purity; Dr. George K. Burgess, director of the Bureau of Standards, Washington, D. C., who kindly had some of the alloys analyzed for oxygen and nitrogen, and Walter M. Saunders and Walter M. Saunders, Jr., official chemists, Providence, Rhode Island, who accurately and efficiently made the chemical analyses of the alloys studied in this paper.

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DISCUSSION

Written Discussion: By O. W. McMullan, metallurgical department, The Timken-Detroit Axle Co., Detroit, Michigan.

The preparation of this paper has involved a great amount of careful work by the investigator and should prove to be of practical value as well as of scientific interest.

The results should be of real value to the user of carburized steel and those of eutectoid composition or higher. Since excess carbides are the hardest and most wear resistant constituent possible to obtain in steel a certain amount of them is desired in gear teeth, bearings, etc., to resist wear and abrasion under heavy loads. It is generally agreed that the higher nickel steels do not have as good wear resisting properties as some of the other common alloys and it is quite probable that its lesser effect on lowering the eutectoid composition has something to do with this. A case of the same carbon content on some of the other steels would have much more excess carbide.

The ability to obtain steels of eutectoid or hypereutectoid composition of lower hardening power as shown by several of the tables suggests the possibility of obtaining steels with better bearing properties than are now available. This would come about through the presence of carbides in a low carbon eutectoid which when hardened would give a softer, tougher martensite. The difference in hardness between the hard particles and softer background could be widely varied. The result would be a tougher, more shock resistant steel with greater freedom from grinding checks and spalling on corners and bearing surfaces.

Author's Closure

The author wishes to express his thanks to Mr. O. W. McMullan for his comments on the research reported in this paper. It is hoped when time permits that the response of some of these eutectoid alloy steels to wear, abrasion

Effectiveness of Special Elements in Displacing the Eutectoid Point in a 0.40% Carbon Steel

Eutectoid Displacement	V	Mn	Cr	Mo	W	Ni
V=2.25%	...	1.44	3.40	2.53	4.00	4.53
Mn=3.25%	0.69	...	2.35	1.75	2.77	3.14
Cr=7.65%	0.29	0.42	...	0.75	1.18	1.33
Mo=5.70%	0.39	0.57	1.34	...	1.58	1.79
W=9.00%	0.25	0.36	0.85	0.63	...	1.11
Ni=10.20%	0.22	0.32	0.75	0.56	0.88	...

Effectiveness of Special Elements in Displacing the Eutectoid Point in a 0.50% Carbon Steel

Eutectoid Displacement	V	Mn	Cr	Mo	W	Ni
V=1.35%	...	2.15	3.11	3.37	5.92	6.44
Mn=2.90%	0.47	...	1.45	1.57	2.76	2.07
Cr=4.20%	0.32	0.69	...	1.08	1.90	2.07
Mo=4.55%	0.30	0.64	0.92	...	1.76	1.91
W=8.00%	0.17	0.36	0.52	0.57	...	1.09
Ni=8.7%	0.16	0.33	0.48	0.52	0.92	...

Effectiveness of Special Elements in Displacing the Eutectoid Point in a 0.60% Carbon Steel

Eutectoid Displacement	V	Mn	Cr	Mo	W	Ni
V = 0.75%	...	3.00	3.13	4.50	8.00	8.67
Mn = 2.25%	0.33	...	1.09	1.51	2.67	2.89
Cr = 2.45%	0.31	0.92	...	1.35	2.45	2.65
Mo = 3.40%	0.22	0.66	0.72	...	1.76	1.91
W = 6.00%	0.13	0.38	0.41	0.57	...	1.08
Ni = 6.50%	0.12	0.35	0.38	0.52	0.92	...

Effectiveness of Special Elements in Displacing the Eutectoid Point in a 0.70% Carbon Steel

Eutectoid Displacement	V	Mn	Cr	Mo	W	Ni
V = 0.50%	...	3.00	3.20	4.50	8.00	8.70
Mn = 1.50%	0.33	...	1.06	1.50	2.67	2.90
Cr = 1.60%	0.31	0.94	...	1.41	2.50	3.34
Mo = 2.25%	0.22	0.67	0.71	...	1.78	1.93
W = 4.00%	0.13	0.38	0.56	0.56	...	1.09
Ni = 4.35%	0.11	0.35	0.37	0.52	0.92	...

and fatigue tests may be studied thoroughly. It is obvious from the impact values of the several alloy eutectoid steels shown in Table XXXV that some of these special steels are tough and also combined with a considerable degree of hardness. From the values of the eutectoid displacements shown in Table XXXVI, the effectiveness of the special elements on the eutectoid displacements have been calculated in the following tables. For example, in the series of "Effectiveness of Special Elements in Displacing the Eutectoid Point in a 0.40 per cent Carbon Steel," as shown in the foregoing, it is necessary to use respectively amounts of manganese, chromium, molybdenum, tungsten and nickel which are 1.44, 3.40, 2.53, 4.00 and 4.53 times that of vanadium in order to produce a 100 per cent eutectoid structure. We must not forget that these figures are based on a study of pure iron-carbon steels to which pure elements have been added.

MICROSCOPIC STUDY OF EFFECTS OF HEAT TREATMENT UPON FERRITE BANDING IN STEEL

By Wilber E. Harvey and Bradley Stoughton

Abstract

This paper reports the results of an investigation as to the possibility of breaking up or eliminating ferrite banding in hypoeutectoid steels. The authors present photomicrographic data showing the results of various heat treatments on this banding.

It was found that, with small specimens, at least, it was possible (1) to almost entirely eliminate ferrite banding by a long heating above the Ac_3 followed by a water quench and subsequent tempering, (2) to entirely eliminate ferrite banding by a long heating above the Ac_3 followed by an air cool (normalizing), (3) to considerably improve banded structures by a long heating above Ac_3 followed by furnace cooling.

Introduction

FERRITE ghosts, a form of fiber found in hypoeutectoid steels, are the remains of an interdendritic segregation from the primary crystallization of the steel in the ingot stage. In the solidification or freezing zone, the early solidifying material, i.e., the dendrites, is of relatively low carbon content. The interdendritic material, having the lower melting point, freezes last and consists initially of higher carbon and high metallic impurity content. Later the position of the carbon is reversed. The cores or dendrites have been enriched in carbon and the interdendritic material has been impoverished in carbon. Upon rolling or forging, the interdendritic material may be fabricated into plates or bands, hence the nomenclature of "banding." If the reduction during rolling or forging is equal along both transverse axes of the ingot, then banding will occur. If the reduction is principally along one axis, then sheets or layers of interdendritic material will result. The ultimate effect of this so-called "banding"
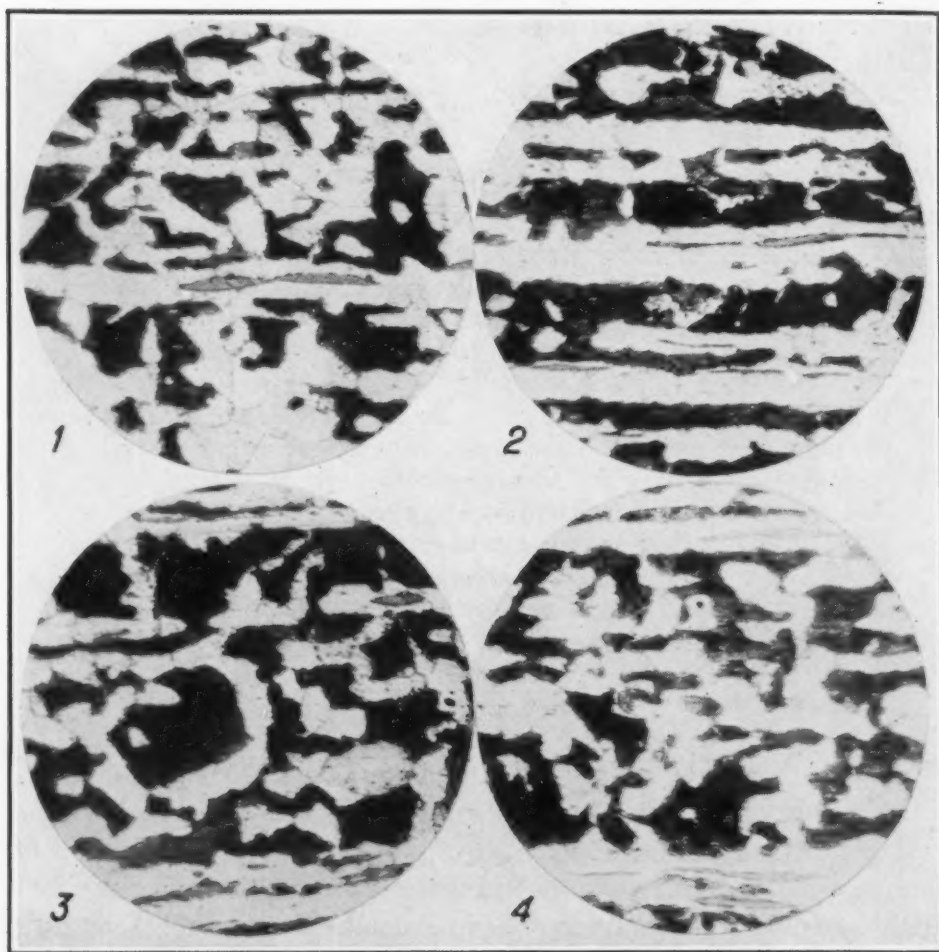
Figs. 1–4—Photomicrographs of the Structure of the Steel Used in this Investigation. Made from Specimens in the As-Rolled Condition. The Original Photomicrographs in this Paper were Made at a Magnification of 100 Diameters but have been Reduced to 90 Diameters in Making Half Tone Reproductions.

is the high ratio of longitudinal physical properties to transverse physical properties, an undesirable feature in most types of steel. These physical properties include tensile strength, ductility and shock resistance.

Thus fiber, while a direct product of the working of the steel, must be dependent upon an initially segregated material or something which will afford an axial distribution of materials. It has been mentioned that the dendrites and interdendritic material are of different chemical composition and this varying chemical composition or segregation may be promoted by:
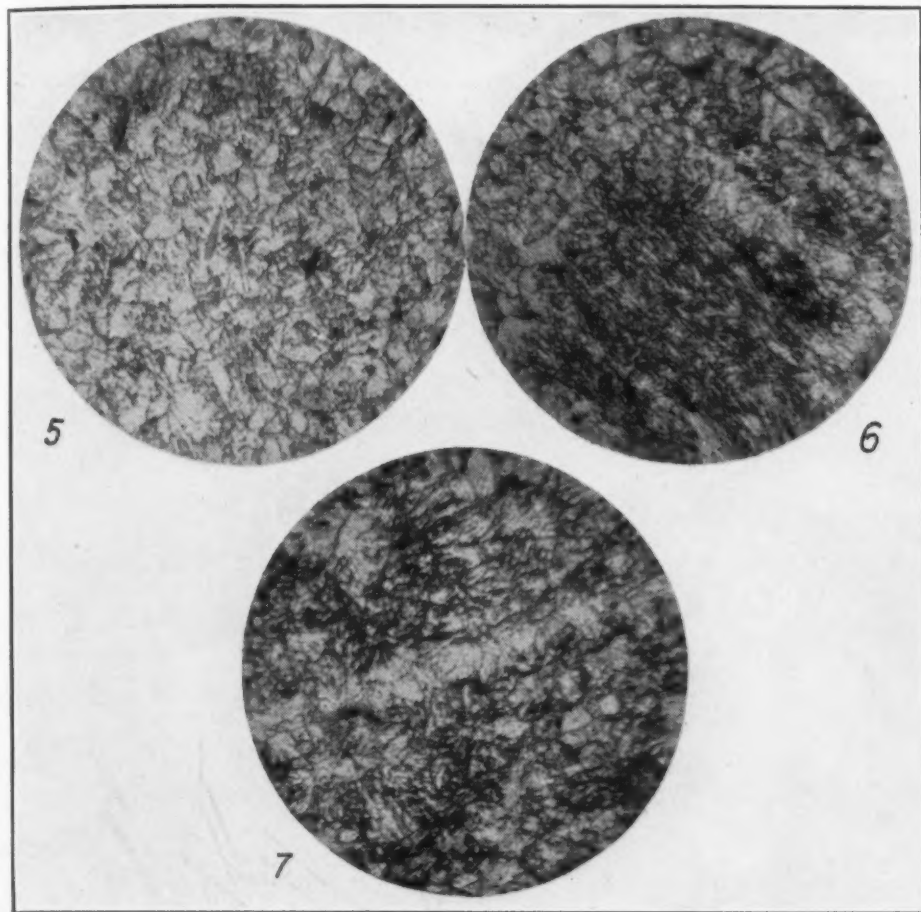
Figs. 5–7—Photomicrographs of the Same Steel as Shown in Figs. 1 to 4 After Quenching from 925 degrees Cent. (1700 degrees Fahr.) and Tempered at 725 degrees Cent. (1340 degrees Fahr.). × 90.

1. High phosphorus content, also possibly a high manganese content
2. High oxygen and nonmetallic content, i.e., "blowholes" and "sonims"
3. Large sizes of dendrites.

Large size, described by Howe as loose packing of dendrites, is primarily an ingot fault which will produce a small number of large dendrites rather than the desirable large number of small dendrites. To effect a large number of small dendrites the following conditions should be promoted:

1. Quiet cooling
   a. Thoroughly deoxidized
2. Rapid solidification
   a. Low casting temperature, to avoid excessive heating of ingot molds before solidification begins in ingot
   b. Thick cold walls
   c. Stripping practice which will produce maximum heat radiation
   d. Small ingots.
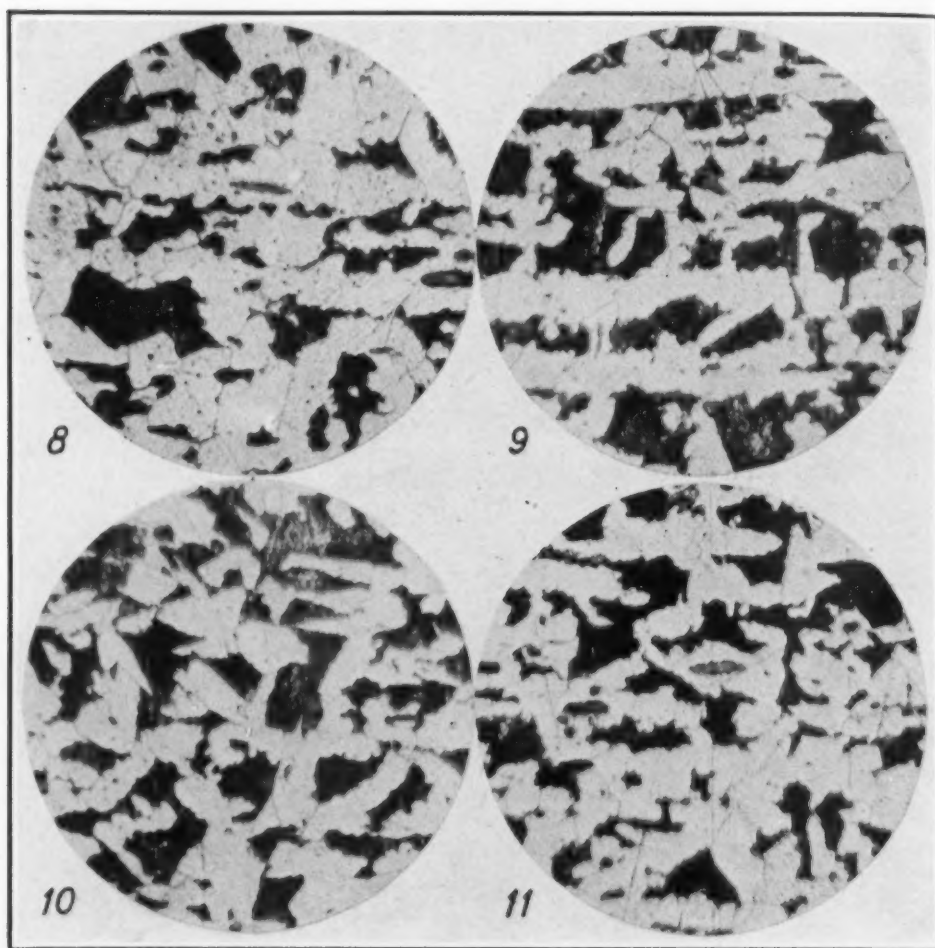
Figs. 8-11. Photomicrographs of the Same Steel as Shown in Figs. 1 to 4 After Long Heating at 925 degrees Cent. (1700 degrees Fahr.) Followed by Furnace Cooling. × 90.

The fiber already present in the steel before working, while broken up during working, may reappear with too high a finishing temperature in working. The closer proximity of the finishing temperature to the critical range, the less danger exists of excessive banding.

Papers have been published concerning the mechanism of the fiber formation and it is not the purpose of this treatise to discuss this problem further.

The paper is a progress report upon a research in which the authors aim to eliminate banding from the fabricated material by heat treatment. There seems to be little, if any, enlightening literature upon this phase of the subject and it is with this idea in mind

that the authors have begun a study of the removal of ferrite ghosts by heat treatment.

PRESENT INVESTIGATION

A perfect study of the problem under investigation would require a complete control of the three main variables governing banding and a separate study of the effects of each variable. It is hoped that this work will be possible at some time in the near future. However, it would seem from practice as if the banding due to ingot conditions and the banding as a result of high phosphorus content provide the majority of serious bandings in steel. It was the fortune of the authors to have at their disposal some badly banded steel of low phosphorus, low oxygen and low nonmetallic inclusion content and this condition afforded a study with seemingly only one variable present—some defect in ingot condition. This paper will be a résumé of several heat treatments used to eliminate banding in this particular steel.

The steel studied was of the following chemical composition:

	Per Cent	Range Per Cent
Carbon	0.20	0.18 to 0.23
Manganese	0.61	0.50 to 0.95
Phosphorus	0.024	0.016 to 0.031
Sulphur	0.041	0.035 to 0.047
Slag and oxides	0.088	
Vacuum Fusion Oxygen[1]	0.002	0.001 to 0.003

The phosphorus, dissolved oxygen, slag and oxide contents are all fairly low and are in small enough quantities to reasonably suspect that they had little if any effect upon the banding of the steel. Thus it would seem as if the banding were due to some mechanism in the cooling of the ingot. The microstructure of the steel as rolled is shown in Figs. 1 to 4 inclusive.

If the banding were due mainly to some ingot condition, then it is logical to expect that it could be removed by a heat treatment. Reheating above the upper critical range, followed by a quench, would be expected to temporarily remove all traces of banding in a steel of this type. This fact has been known for many years. However, in past experiments, the banding reappeared upon reheating the quenched material below the critical range.

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[1]The authors are indebted to Doctor Anson Hayes of the American Rolling Mill Co. for his kind assistance in making the vacuum fusion oxygen determination.
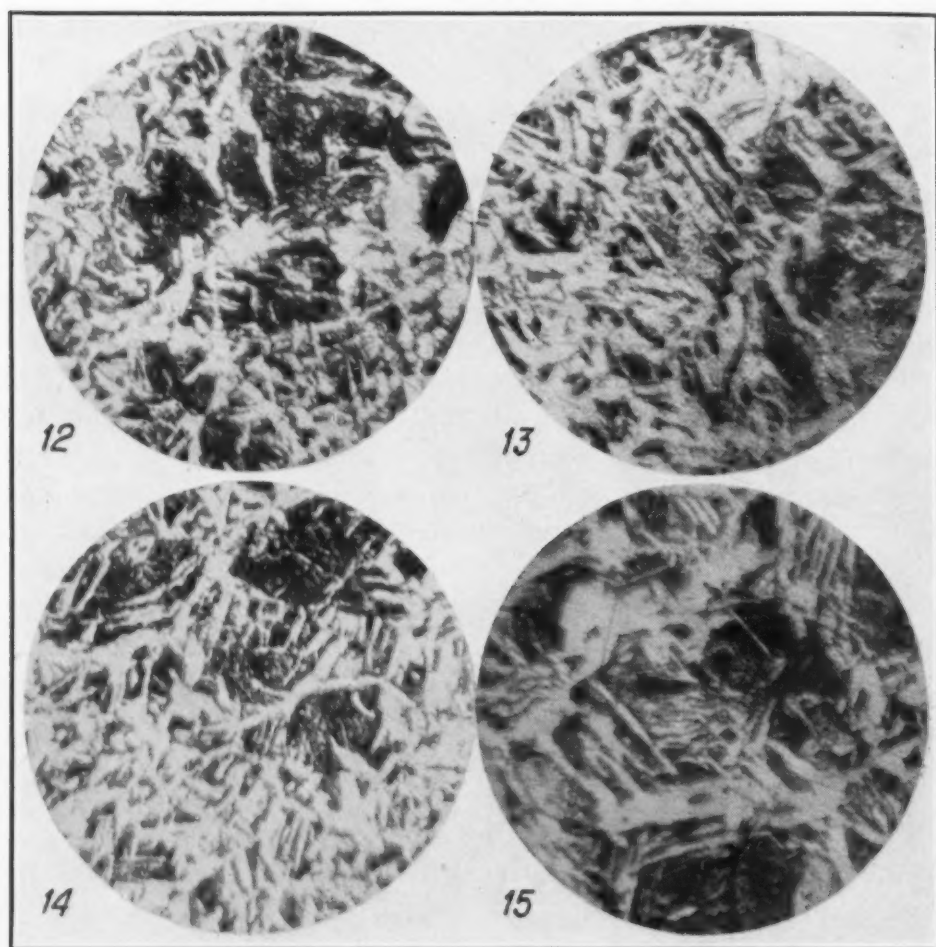
Figs. 12-15—Photomicrographs of the Same Steel as Shown in Figs. 1 to 4 After Long Heating at 925 degrees Cent. (1700 degrees Fahr.) (Normalizing) Followed by a Still Air Quench. × 90.

By subjecting the steel under study to a long soak at fifty degrees above the critical temperature, the banding was eliminated after quenching and the banding remained eliminated after a 3-hour temper at 930 degrees Fahr. (500 degrees Cent.). A sample of the quenched steel was then tempered at 1290 to 1335 degrees Fahr. (700 to 725 degrees Cent.), for four hours and the banding remained eliminated with the exception of a few slight bands. Photomicrographs of the quenched steel tempered at 1290 degrees Fahr. (700 degrees Cent.) are shown in Figs. 5 to 7 inclusive. Figs. 5 and 6 show the average microstructure of the piece, while Fig. 7 shows the most pronounced band in the specimen. The position of the nonmetallic inclusions in Figs. 5 and 6 show their removal from a

strictly ferrite area into a mixed ferrite and carbide area of the steel.

Thus in a fabricated part which is of such design that a quench and tempering heat treatment can be used, a long heating above the upper critical range, followed by a quench and a temper, affords a fairly adequate means of removing ferrite banding in hypoeutectoid steel. This is evidenced by a study of Figs. 5 to 7.

Many large fabricated parts are not capable of being quenched, their large sizes being prohibitive of such a heat treatment. Still other parts of intricate design cannot be quenched without distortion. With this thought in mind, a long heating 90 degrees Fahr. (50 degrees Cent.) above the critical range followed by a furnace cooling was tried and the results are shown in Figs. 8 to 11 inclusive. The banding is fairly well broken up although there still remain considerable traces of the original banding. A similar heat treatment followed by a still air quench was the next heat treatment attempted and this normalizing treatment revealed a complete removal of banding. Photomicrographs of this structure are shown in Figs. 12 to 15 inclusive. The normalizing treatment offered the most complete removal of the banded structure.

Details of the different heat treatments are as follows:

1. Heated to 1695 degrees Fahr. (925 degrees Cent.), held three hours, quenched in cold water, reheated to 1290 degrees Fahr. (700 degrees Cent.), cooled in air, Figs. 5 to 7 inclusive
2. Heated to 1695 degrees Fahr. (925 degrees Cent.), held three hours, furnace cooled, Figs. 8 to 11 inclusive
3. Heated to 1695 degrees Fahr. (925 degrees Cent.), held four hours, air cooled, Figs. 12 to 15 inclusive.

The grain size revealed in all of the photomicrographs is rather large. This is due both to the high temperature, approximately 90 degrees Fahr. (50 degrees Cent.) above the critical range, and also the unusually long sojourn at that temperature. The specimens treated were all very small in size, thus adding to the effect of the heat treatment. Although the specimens studied were of small size, the results obtained were of enough significance to warrant the belief that similar heat treatment would produce an effective result on large pieces. It was impossible to attempt heat treatment of large pieces with the facilities available.

The next series of tests by the authors will represent an attempt to remove that type of banding which is caused by high phosphorus content and the results of these tests will be reported at an early date.

Conclusions

Hypoeutectoid ferrite banding which is due to faulty ingot conditions can, at least in small size specimens, be:

1. Almost entirely eliminated by a long heating above the Ac₃ followed by a water quench and any tempering treatment not higher than 1335 degrees Fahr. (725 degrees Cent.)
2. Entirely eliminated by a long heating above Ac₃, followed by an air cool, i.e., a normalizing treatment
3. Considerably bettered by a long heating above Ac₃ followed by furnace cooling.

DISCUSSION

Written Discussion: By Herbert M. Boylston, Professor of Metallurgy, Case School of Applied Science, Cleveland.

Any new information upon the subject of ferrite banding in steel is always welcome and excites our interest. This paper by Harvey and Stoughton is no exception. For many years we have been acquainted in a general way with the relation between banding and directional physical properties in steel, but quantitative measurements of the importance of banding as a defect in steel are sadly lacking. We really do not know how much banding is allowable in good quality steel. I understand that Dr. C. H. Herty, Jr., and his associates are planning to carry out some investigations on the relation between nonmetallic inclusions and the physical properties of steel. We at Case School have recently been greatly encouraged in graduate work by the acquisition of several graduate laboratory rooms and a considerable gift of high-grade apparatus and expect to have seven or eight graduate students in metallurgy this fall and hope to do some work connected with the question of banding in steel.

We must all admit that there are many commercial steels which are more or less badly banded. I sometimes think that there are few if any rolled steels which are entirely free from this defect although the degree of banding varies greatly in different steels. If there is or can be developed a method of removing banding by any comparatively simple heat treatment such as normalizing we should know it and practice it frequently, but after reading the present paper carefully, the question quickly asserts itself,—have the conclusions been reached a bit hastily?

Composition—It does not seem at all certain that 0.024 per cent phosphorus content is low enough to eliminate it as a suspect, for, through local or dendritic segregation, the phosphorus, in my opinion, might easily run as high as 0.04 or 0.045 in the ferrite bands. Sulphides or other nonmetallic inclusions occupy their usual places in the ferrite bands in the steel studied by the authors and no doubt exert their influence.

Etching Reagents—Although not stated in the paper, it would seem from inspection of their illustrations that they used either picric acid or nitric acid as an etching reagent. It has been well-known for years that a much better reagent for detecting banding is Stead's reagent or Le Chatelier's reagent, both of which are solutions of copper chloride which deposit copper in different de-

grees upon the surface of the sample depending upon the local analysis of the latter.

To illustrate the difference produced by Stead's reagent as compared with nital, I would refer you to two which I have just received from one of the Sauveur and Boylston Correspondence Course students, M. L. Samuels, of Birmingham, Ala. The steel used was the same in both cases and consisted
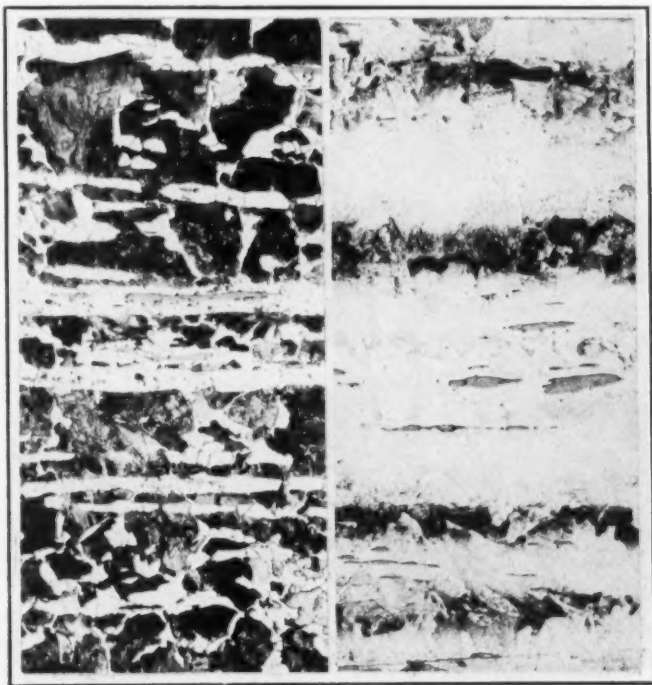

Fig. 1 (Left)—Rolled Rail Steel Etched in Nital Showing Ferrite Streaks. X100. Fig. 2 (Right)—Same Specimen as Fig. 1 Etched One Minute in Steads Reagent. X 100.

of a section of an old steel rail which I knew to be badly banded. I suspect that if the authors of the present paper had etched their specimens with Stead's reagent, that even their normalized specimens would have shown a considerable amount of banding due to dendritic segregation, even if the actual ferrite bands as shown by etching with picral or nital have nearly disappeared. I suggest that they try this.

Physical Properties—Since we are interested even more in the directional physical properties of banded steels than in their microstructure, it would be interesting to know if the authors convinced themselves by such tests as to whether the *effects* of banding have been eliminated by their normalizing treatment.

Grain Size—If a structure of such coarse grain size as shown in their Figs. 12-15 is a necessary corollary of any attempt (which appears at all successful) to remove banding, would they care to recommend such a steel for structural use? In other words, is the cure worse than the disease?

Written Discussion: By Albert Sauveur, Gordon McKay professor of metallurgy and metallography, Harvard University, Cambridge, Mass.

The possibility, if it exists, of removing "ghosts" and banding by some commercially practical treatment would undoubtedly be warmly received by steel makers. The authors report that in the case of low carbon, low phosphorus steel at least it can be done. They state, however, that the treatments recommended have been applied only to "hypoeutectoid ferrite banding which is due to faulty ingot conditions." As I understand it, this faulty ingot condition results from the production of large dendrites during solidification. While it may be true that large dendrites should result in larger bands or ghosts, it does not seem as if there should be a sharp distinction between the behavior of steel made up of large dendrites and one made up of small dendrites when subjected to identical treatments.

The steel studied contained 0.024 per cent phosphorus and was "badly banded." From the low percentage of phosphorus present, they infer that this banding must have been caused by some "defect in ingot condition." In my study of dendritic segregation and resulting banding, however, I have repeatedly found these characteristics in steel containing 0.02 per cent phosphorus and even less. As some of these steels had been prepared by melting less than one pound of metal in crucibles, the dendritic segregation could not be ascribed to "ingot conditions." In a paper presented at this meeting by Dr. Reed and myself, we show the marked dendritic segregation existing in a small ingot containing but 0.004 per cent phosphorus, 1.06 per cent carbon and substantially free from manganese, sulphur and silicon. As dendritic segregation can hardly in this case be ascribed to phosphorus, it is concluded that carbon is capable of producing this phenomenon.

It is obvious that in order to erase dendritic segregation it is essential that homogeneity in chemical composition be brought about between the axes and fillings of the dendrites existing in castings or between the so called "ferrite bands" and surrounding metal in forgings. To promote this chemical homogeneity, we naturally turn our attention to the action of a long sojourn at high temperature where diffusion should be permitted to exert itself. It is believed that the carbon diffuses readily while phosphorus does so very reluctantly. On slow cooling, the action of phosphorus will again force the carbon out of the position it (the phosphorus) occupies and the dendritic pattern or banded structure will form again. Relatively fast cooling, at least, appears therefore to be necessary, when phosphorus is present, to retain the homogeneous distribution of the carbon existing at high temperatures. This is confirmed by the results reported by the authors. In the substantial absence of phosphorus, the removal of banding should be more easily brought about.

The results of the experiments to be conducted by the authors will be awaited with much interest.

Written Discussion: By L. E. Ekholm, metallurgist, Harrisburg Pipe and Pipe Bending Co., Harrisburg, Pa.

The writer greatly appreciates the work done by the Messrs. Harvey and Stoughton on the elimination of banded structures. This condition is, at times, very serious and sometimes necessitates a great deal of work by a forge shop

to break it up. Any simple method of elimination of this so called banding is greatly appreciated.

The fundamental aim of the elimination of these so called bands is to reduce to a minimum the high ratio of longitudinal to transverse physical properties. Microscopic record shows that the authors have eliminated in some cases and partially eliminated in other cases this banding. The question is just how much the ratio between longitudinal and transverse properties has been reduced. The writer has noted on numerous occasions where a banded structure has, as shown by microscopic examination, been apparently eliminated by high normalizing temperatures or almost fully eliminated as was the case in the paper under discussion, but where the marked differences in directional properties still persisted. The variations in directional properties were found both in tests and in forgings in service. The data gathered was from a wide range of analysis, ranging from straight carbon to higher alloy steels. The variations showed up in ductility, in tension, and in impact values.

The writer would appreciate any information on ductility figures that is available which shows how efficiently the variations in directional properties can be eliminated by heat treatment.

Written Discussion: By C. R. Wohrman, in charge of research, General Plate Co., Attleboro, Mass.

The problem taken up by Stoughton and Harvey is one of great practical as well as theoretical interest, and it is gratifying to learn that further work is contemplated on this problem.

It is perhaps timely to suggest that the microscopic studies be supplemented by physical tests from which the extent of any improvement resulting from the apparent elimination of banding could be evaluated.

Dendritic structures as revealed by etching reagents cannot be taken to serve as a true measure of the quality of the steel, it having been time and again observed that steels of most excellent physical properties can be made to yield just as beautiful dendritic pictures as low grade steel. This, of course, applies to banded structures resulting from the deformation of dendrites as well. The pictures are an effect obtainable, as it were, from a large and a small cause alike, the cause here being chemical heterogeneity of different degrees of seriousness.

The authors describe this heterogeneity as being promoted by three different agencies of which the one taken up in the present paper is termed "large size of dendrites." Large size dendrites on deformation are bound to yield large size banding, and one wonders why a magnification of 90 had to be used by the authors to show this banding when ordinary dendritic structures making no claims of being large can be observed at quite low magnifications. We note in this connection that macro-etching with Le Chatelier's reagent commonly employed for the detection of dendritic segregation and banding resulting therefrom has apparently not been employed by the authors. The etch used is not indicated but it seems to be a nital etch which differentiates perfectly between pearlite and ferrite but is not suitable for showing any segregation in the ferrite. If the ferrite bands shown here are due to dendritic segregation (and in that case represent elongated dendritic fillings) we must conclude that these

fillings were sufficiently rich in phosphorus or other impurities to expel the carbon into the axes, which would bring us back to the other two main variables (phosphorus content, oxygen and nonmetallic content) which the authors claimed eliminated. It would seem in other words that large size dendrites or faulty ingot conditions per se cannot be made to account for the banding observed, except insofar as they promote the segregation of phosphorus and other impurities.

As to the elimination of banding the difference between complete or permanent elimination and partial or temporary elimination should be pointed out. The former means complete homogeneization of the metal, the latter merely a temporary elimination of carbon segregation. Soaking at high temperatures causes the carbon to be dissolved in the iron and to become uniformly distributed. If quickly cooled (as by quenching or air cooling of a small specimen) the carbon is prevented from segregating away from the impure portions and a state of fair homogeneity in carbon distribution is achieved. This is of course all that is wanted in many cases, and all that appears to have been accomplished by the heat treatment suggested by the authors. This treatment seems inadequate to effect permanent homogeneization. The authors' Figs. 8 to 11 pertaining to treated steel slowly cooled resemble very closely the picture of the original untreated steel (Fig. 1 to 4) except in that they show more ferrite. This I take to be due to decarburization effected during the long heating.

Written Discussion: By C. H. Herty, Jr., director of research Mining and Metallurgical Boards, Carnegie Institute of Technology, Pittsburgh.

The authors are to be congratulated on the heat treatment worked out for the correction of banding in the steel studied. It appears to me, however, that the statements made as to the cause of the banding are not entirely consistent and it might therefore be possible that this particular heat treatment would not be general in nature.

First of all, it is stated that hypoeutectoid ferrite banding which is due to faulty ingot conditions can be corrected by the heat treatment given. This term "faulty ingot conditions" is far-reaching. It is stated on page 181 that the steel studied was of low phosphorus, low oxygen, and low nonmetallic inclusion content, whereas the actual composition is 0.024 per cent phosphorus, 0.041 per cent sulphur, and the oxygen and slag and oxide contents do not compare with each other in any way. It is not stated how the slag and oxides were determined, but in view of the low vacuum fusion oxygen content it would appear that aluminum must have been used in the deoxidation of this steel. I hardly believe that this could be called a low phosphorus, low sulphur steel, and from Figs. 1 to 4 it would certainly appear that the ferrite banding was at least accompanied by sulphide streaks.

On page 177 the statement is made that the interdendritic material consists initially of higher carbon and high metallic impurity content and that later the position of the carbon is reversed. This is a point which is difficult to visualize—that is, the diffusion of carbon from a high to low concentration and finally from a low to a high concentration. This does not seem possible.

Oral Discussion

C. H. HERTY, JR.: After we got Professor Stoughton's paper, we began

to do some work of our own along those lines on some very badly banded steel and some that was only slightly banded. So far we have found that Professor Stoughton's treatment checks right up. We have found that we can predict quite nicely what time of normalizing is required, and on the worst banded steel we have found that an hour and a half normalize is just as good as a three-hour normalize to relieve that banding.

YAP, CHU-PHAY:[1] I had not intended to discuss this paper because I thought there would be others better qualified than I to discuss it. I have been quite interested in banding and have done considerable work on it.

It may be heretical, however, of me to say that I do not at all believe banding is entirely due to phosphorus segregation.

About four years ago I submitted a paper to the American Institute of Mining Engineers, which was accepted provided I rewrite some parts of it, but I never took the trouble to do so. I showed that starting with a low carbon steel of about 0.10 per cent, which absolutely did not show banding in any form, and by reduction in cross section I was able to produce artificial banding which could not be destroyed, no matter what kind of treatment was given to it.

I believe the effect of work strain is much more persistent than what we believe it to be. I also believe that cold working, very severe cold working as in strips, does impress upon the steel certain directional strains which may cause the banding. I think the banding, at least in low carbon steels, should be expressed not so much in terms of the carbide as a term of ferrite, that is, the position of the carbide is entirely conditioned by the distribution of the ferrite.

My experience also indicates that although banding can apparently be removed by normalizing, it seems that banding comes back after a reanneal, even after three hours' normalizing; so I do not know whether Professor Stoughton has verified this point.

J. FLETCHER HARPER:[2] I would like to ask Professor Stoughton if, after his normalizing treatment or the quenching treatment, he reheated the piece above the upper critical temperature range? And if after such reheating if the banding did not re-occur?

In a number of experiments we have conducted along the same line we have always found that the banding re-occurs upon heating above the Ac_3 critical range. In our experience with forged products we have found that the banded condition can often be produced in the forging operation and that it is somewhat questionable whether it can all be blamed upon the ingot condition.

We have further found that banding which has originally taken place in the direction of forging elongation, can be broken up by an upsetting operation in the opposite direction; this work being done at a proper temperature.

We have sometimes questioned whether or not the presence of inclusions and the effect of their rate of cooling upon the adjacent material does not have something to do with the segregated ferrite or banded ferrite condition.

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[1]Consulting physical metallurgist and chemist, 27 Grove St., New York City.

[2]Chief research engineer, Allis-Chalmers Manufacturing Co., Milwaukee.

O. W. ELLIS:[3] I was particularly interested when this paper came to hand because a few weeks before it arrived I had had occasion to investigate the cause of failure of a gear, which, when examined, was found to show marked banding on being etched. The fact that this banding showed up so clearly made us feel that here was an interesting piece of steel to investigate. The steel was a 0.23 per cent carbon, 3 to 3.5 per cent nickel steel. We took a series of 8 samples from this steel, determined its critical ranges, and then heated one of the samples to a point just beyond the critical range, another to 100 degrees higher, another to 200 degrees higher, and another to 300 degrees higher, allowing them all to cool in air. The samples were small, so that the rate of cooling was fast. Another group, similarly treated, was quenched in water. In every case the half-hour treatment eliminated the banded structure as far as the normalized samples were concerned, despite the fact that we quite expected to discover evidence of banding. The point has already been made in the discussion that the removal of ferrite banding does not of necessity mean that all banding has been eliminated, as can be made quite clear by etching with such a reagent as Stead's. It was found that all the samples referred to above showed distinct banding on etching with Stead's reagent.

I would not have risen to speak, had it not been that we became so interested in this question that we started a series of experiments on the same subject, so that it is possible that about the same time that Professor Stoughton's paper is published we may have something similar to present. It will be interesting to see how our two series of experiments compare.

Authors' Closure

It is always a pleasure to an author to have any publication discussed by able contemporaries and there is no exception to the rule in the case of this paper. It is exceedingly pleasing to find such able metallurgists as Professors Boylston and Sauveur discussing our paper.

Throughout Prof. Boylston's discussion of our paper it would seem as if, perhaps at times, the suggestions offered by Prof. Boylston are irrelevant to the problem studied. This is particularly true in the suggestion that Le Chatelier's and Stead's reagents be used as etchants for the specimens in which it was claimed that the ferrite banding had been removed. The authors claim to have removed ferrite banding only from the steel by heat treatment and do not claim to have altered any other banding in the specimen, i. e., phosphorus banding, etc. As a matter of fact, it is known that the phosphorus did occur in bands—such evidence having been proven by an etching with Stead's reagent. That the phosphorus banding had little or nothing to do with the ferrite banding was easily determined. In a normalized specimen, the phosphorus banding, with the use of Stead's reagent, was shown to be in perfectly straight lines. The distribution of the ferrite in the specimens, normalized to remove ferrite banding, is very irregularly distributed in grain boundaries and cleavage planes, showing no orientation to the phosphorus bands whatever.

The answer to Prof. Boylston's question as to whether the grain size as shown in Figs 12-15 is a necessary corollary of any attempt to remove band-

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[3]Director of metallurgical research, Ontario Research Foundation, Toronto, Canada.

ing, I refer to page 183 where it is suggested that the coarse grain size is due to first a fairly high temperature, secondly the length of sojourn at that temperature, and finally the small size of the specimens used. In one specimen, tested since the photographs were made, very good grain size has been obtained with the normalizing treatment. The cure is certainly not worse than the disease.

The authors regret that they have no physical properties to correlate with the micrographic examination. The only comment possible is that the fractures in the specimens treated for the removal of ferrite banding are considerably better in appearance than the fractures in the banded material.

Professor Sauveur makes the statement, "While it may be true that large dendrites should result in larger bands or ghosts, it does not seem as if this should be a sharp distinction between the behavior of steel made up of large dendrites and one made up of small dendrites when subjected to identical treatments." While we agree with Professor Sauveur that there is no sharp distinction between the behavior of steel in larger bands or smaller bands, nevertheless, we think he will agree that the larger the bands, the more difficult it may be to eliminate them.

The subject of the influence of the phosphorus upon the ferrite banding, we believe, has already been discussed in answer to Professor Boylston's suggestions and it is applicable to Professor Sauveur's discussion.

Regarding Dr. Herty's question about the statement on page 177. That statement was made in an attempt to express what is in the text book on this subject and what the chairman has said so very well himself; namely, that the segregate will be high in phosphorus and in carbon but when the carbon has an opportunity to diffuse, the phosphorus forces the carbon out of that into the other part. Perhaps we have not stated it as well as we should, but it is that well known fact that we intended to express. We had no expectation of precipitating discussion of the causes of banding, or the rôle of phosphorus, or how little phosphorus would cause it. We simply took what was well established in the literature as to the causes of banding and put them down. We simply wished to say that with 0.024 per cent phosphorus, we have shown certain results in eliminating the microscopic banding. We do not wish to say that with higher phosphorus we would do so. So when we say that there was not enough phosphorus to cause the banding, we mean that we do not claim that the treatment that we have given would eliminate banding if there were more phosphorus to cause banding, or, if there was more severe banding. Perhaps that expresses a little more clearly what we had in mind.

Answering Mr. Harper, we did not heat any of the normalized pieces above the critical temperature after normalizing. We did heat treat them to a high point, as stated in the paper, and allow them to cool in the furnace, and we heated some of the normalized pieces to the tempering temperatures, which are customary below the critical point but not above. That may be a very interesting thing to do.

I want to say that our experience also checked that of the last speaker, that in one case we etched with Stead's reagent and found segregation of phosphorus, and still we did not have microscopic ferrite banding, so it is ad-

mitted that there is banding even though there may not be a ferrite banding that shows up under the microscope.

In answer to Mr. Wohrman's question as to why a magnification of 90 diameters has been used to show the banding when magnifications of much lower power would have revealed the banding just as well, it is true that a lower magnification would easily have revealed the banding, which is decidedly marked. But when most of the banding had been removed, such as had been done by our heat treatment, it would have been rather difficult to even find the traces of the banding which remained at such low magnifications. It was with an idea of comparing the remnants of the banding with the original size of the banding that a magnification of 90 diameters was chosen. Mr. Wohrman suggests that Le Chatelier's reagent was not used by the authors in the determination of the banding in the steel. This is quite true and the authors were well aware of this fact. Le Chatelier's solutions are commercially used for the segregation study and are not particularly adapted to the banding of ferrite itself. Again the title of this paper should be emphasized—that the study was one upon the ferrite banding in steel and not one upon the distribution of phosphorus in the steel. The authors do not claim in any way to have removed any phosphorus which might have been segregated in the steel but only to have moved the relative positions of carbon and ferrite.

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